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STEEL FOUNDARY PRACTICE

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STEEL FOUNDRY
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CHAPTER I

PROPERTIES OF STEEL CASTINGS

1. PRODUCTS OF STEEL FOUNDRY

The combination of high strength and ductility makes steel much superior to other alloys and nonmetallic construction materials. Steel is highly resistant to stresses produced by variable and impact loads. Mechanical structures and assemblies made of alloy steel can be used at low (—100° C) and high (not above 700°C) temperatures. Nickel, cobalt, molibdenum and tungsten base alloys designed to operate at high temperatures are extremely heat resistant. Parts made by the method of powder metallurgy from refractory carbides of silicon, chromium, etc., can sustain still higher temperatures—above 1500 °C.

Large structures are produced from reinforced concrete in which steel bars form a rigid frame to take over the tensile stresses while the concrete filling operates for compression.

Of late, organic substances called plastics began to supersede steel in many of its applications owing to the progress achieved in chemistry.

Steel ingots are manufactured by rolling, smith and drop forging, press working and casting. Steel castings, such as machine frames, cylinders with shaped interiors, gears, sheaves, cross-beams and many others, have an extremely complex geometry. Manufacture of intricate shapes by casting has the advantage of a lower cost as compared to other methods, for example, forging, which requires subsequent machining. The dimensions of steel castings are always nearer to those of the ready product than those of a forged piece or rolled stock. The highest dimensional accuracy is provided by pressure-die casting, shell moulding and also by the investment pattern method, in which the pattern is melted and burned out, or removed by a solvent. These progressive methods are acquiring an ever greater significance with each passing year.

.Heavy steel casting (up to 200-300 tons), such as anvil blocks for forging hammers, supporting plates of hydraulic presses, etc., cannot be forged or press-formed. Forging methods are inapplicable to them and they can therefore be manufactured only by casting, although a forged product would have a higher and more uniform strength over the entire volume. Still heavier blanks, such as cross-beams of powerful presses, are obtained by welding rolled bars.

Sometimes large blanks are cast in parts and then welded together. Such structures are called cast-weld. Parts of the casing of a steam or gas turbine, for instance, can also be welded together. However, before the one-piece casting method is renounced, the cast-weld construction should be checked for adequate effectiveness.

I. Bobrov, K. Smirnova and the author have proved that a very dense structure of steel can be obtained by solidification under high pressure used to produce car wheels, hubs, gears and caterpillar rollers without any risering and with the minimum machining allowance. The yield of good metal as compared to all molten metal assumed at 100 per cent amounted in the experiments to about 93-95 per cent which is quite unattainable in the production of small parts by forging, press working or casting (25-30 per cent) and in lot production of steel castings including castings made by the precision casting method (50-75 per cent). The high economic and production efficiency of high-pressure solidification of liquid alloys, in general, and of steel products, in particular, makes this method a highly promising proposition.

Pressure-die casting differs from the above method in that it requires a system of risers and gates. High-pressure solidification at 500-700 atm (gauge) considerably increases the yield of good dense metal without any risering. Centrifugal casting of steel pipes and bushings which requires neither risers nor gates successfully competes with the solidification under high pressure.

2. CLASSES OF STEEL CASTINGS

Castings are classified according to their purpose and mode of manufacture.

By their purpose the castings are classed on the basis of steel quality: general-purpose structural, heat-resistant, acid-resistant, alkali-resistant, wear-resistant, etc.

General-purpose castings are most widespread. They are generally made from carbon steel. As to the annual output by weight the first place belongs to steel castings for the tractor building industry, the second for automotive and transport industry, the third for heavy engineering and, finally, for agricultural and general machine-building.

Heat-resistant castings manufactured from high-alloy steel are employed at high temperatures (steam and gas turbines, jet aircraft, oil-refining plant, nuclear equipment, etc.).

Castings made from acid- and alkali-resistant alloys find application in the chemical industry. Wear-resistant castings made of alloy steel are extremely versatile in use; they are employed for

caterpillar tracks, teeth and lips of bucket excavators, scoops of washing machines in hydrometallurgy, tramway spiders, parts of railway switches, side plates of crushers and pulverisers, etc.

By the mode of manufacture the castings are classed as mass, large- and small-lot, individual and unique castings ranging from several dozen to two hundred tons in weight.

The highest accuracy (4-5th grade) for tolerances and allowances distinguishes steel castings produced by the pressure-die method.

The second place belongs to castings made by investment moulding (4 to 7th grades of accuracy). Gravity-die castings come in third (5 to 7th grades of accuracy). Shell moulding is fourth and sand moulding fifth (7 to 8th grades of accuracy).

Steel castings are differentiated by the name of steel from which they are made.

About 75 per cent of all castings are manufactured from carbon steel and 25 per cent from alloy steel which exhibits an increased strength at a high temperature or has special chemical and physical properties.

The carbon content classifies steels into low-, medium- and high-carbon grades. Most of the machine elements are cast from medium-carbon steel which contains from 0.20 to 0.40 per cent carbon. Low-carbon steel having 0.05 to 0.08 per cent C is used in electrical engineering (frames of electric motors, pole shoes, etc.). High-carbon pearlite-ferrite steel (0.5-0.7 per cent C), pearlite or pearlite-cementite and pearlite-graphite steels are used to produce steel rolls or parts operating in conditions of dry friction: jaws of stone crushers, pins and armour of hammer mills and ball mills for coal, clinker, chamotte, quartz, etc.

The share of alloy steel castings has been steadily increasing during the last twenty or thirty years, resulting in a reduced weight of machines and longer life. The production of such castings continues today at a still faster rate.

The content of one or several alloying elements (in per cent) classes steels as

low-alloy up . . . . to 2.5
medium-alloy . . . . 2.5 to 10
high-alloy . . . . . over 10.

Low-alloy structural steel possesses high strength and ductility and a high resistance to impact loads. It can therefore be classified according to its strength.

Medium-alloy steel in small amounts goes to make especially durable tools. Castings made from this steel have an extremely poor machinability.

High-alloy steel is employed to make castings with special chemical or physical properties, such as high resistance to heat, corrosion and wear in a chemically active medium—acids, alkalis, in sea, mine or fresh water, in the atmosphere of flue gases charged with disulphide, in humid air (in the latter case the steel is called stainless).

Castings made of high-alloy steel are intended to operate at high temperatures but under a small specific mechanical load as, for example, hoods of heating furnaces, stack valves and other similar parts manufactured from scale- and heat-resistant steels.

Scale-resistant steel is highly corrosion-proof and heat-resistant steel is practically unaffected by elevated working temperatures.

According to its microstructure, high-alloy steels are subdivided into pearlitic, martensitic, ledeburitic and ferritic classes. In the case of high wear due to abrasion use is made of high-alloy austenitic steel which contains up to 12-14 per cent manganese. It is used for tramway spiders, caterpillar tracks, etc.

High-alloy permanent magnet and nonmagnetic steel goes for the manufacture of such parts as the washers of magnetic cranes.

3. CASTING METHODS

The casting methods depend on the size, weight, quantity and chemical composition of steel and the requirements the cast metal should meet. Irrespective of their size, the majority of castings are made in green or dry sand loam moulds with wooden or metal patterns. In serial production as, for example, at tractor, automobile, railway car and other plants wide use is made of machine moulding with metal patterns. Large and bulky castings are moulded in sand, loam and brick. The moulding sand is packed by pneumatic rammers. Large moulds, up to 40 tons in weight, are packed with jolting moulding machines with the table turning through 180° about its horizontal axis of symmetry, together with the pattern plate and moulding box fixed to the table. Sometimes large moulds are rammed in by powerful sandslingers.

Metal moulds are used rather rarely for steel castings because of their poor strength and high cost of manufacture.

Higher accuracy of steel castings is of prime economic importance. As a rule, the machining allowance and, especially, shrinkage allowance of steel castings are larger than in iron castings. The accuracy of castings, for example turbine blades, has been increased by providing a negative allowance for the size of metal patterns. Many heavy engineering works have increased the accuracy of steel castings by shifting over to moulding in rapidly solidifying mixtures with soluble glass by the detachable pattern parts with a negative allowance.

4. STEEL CASTING PROCESS

The mode of manufacture of steel castings depends on the properties of molten steel (physical, chemical and processing) which form the subject of study in this textbook.

Steel is a refractory alloy. Its melting point is higher than that of pig iron and many nonferrous alloys. The moulds are filled at a temperature which exceeds the liquidus temperature, i. e., with a reserve of heat expended on losses due to radiation and heating the mould walls during pouring. The melting point of some high-temperature alloys (for example, cobalt, nickel, tungsten and molybdenum base, etc.) employed in foundry practice for the manufacture of heat-resistant products (vanes and impellers for jet engines and for other purposes) is higher than the melting point of ordinary steel.

The high temperature of molten steel (1570-1650° C) requires especially refractory materials for foundry moulds which should be protected against premature erosion and corrosion.

The moulding sand mixtures for steel castings should be highly refractory. Loams and sands used for pig iron and nonferrous metal castings are unsuitable for steel and, for this reason, it would be ill advised to make steel and pig iron or steel and nonferrous metal castings in one and the same shop.

Steel is easily oxidised in the air. Therefore, liquid metal has to be frequently protected from the atmospheric oxygen or a reducing medium has to be provided in the mould cavity by adding carbon-containing substances, fluorine salts and boron into the mould washes as in the case of magnesium casting.

Steel has a lower fluidity than pig iron or nonferrous alloys. For this reason, the metal for heavy castings should be fed over the shortest route possible and for large-size castings through a number of gates, with the mould well heated.

While designing foundry moulds for steel castings, the heat content of steel during pouring, heat transfer to various parts of the mould and their heating should be the prime consideration. The heat distribution in molten steel is the principal factor for obtaining a sound casting.

During freezing, the steel is liable to contract by 1.5 to 3.0 per cent. Hence, the shrinkage allowance for steel castings should be larger than for pig iron or nonferrous castings and, for this reason, the patterns used for steel castings are made larger in size than for other alloys. A pattern intended for an iron casting can never be used to make a steel casting.

The volumetric shrinkage of steel averages between 4.5 and 9 per cent. Large concentrated holes in the casting are prevented by risers which are subsequently cut off from the casting. This involves a loss of metal and reduces the yield of good ingot. When

the casting is not required to be very dense, the risers are dispensed with as, for example, in casting coupler heads.

Besides concentrated blowholes, castings may also develop shrinkage porosity. To avoid this, the production of the critical castings must be tackled with great foresight and attention providing for proper shrinkage allowances, correct design of risers, chills and gates and using the proper temperature and the pouring rate.

Large linear and volumetric shrinkages cause the formation of hot cracking, so called to distinguish them from cold or annealing cracks which develop as a result of an unfavourable variation in the temperature in the casting being cooled and the internal stresses which grow in proportion to the temperature difference.

Casting cracks present in sheaves, wheels, gears and other similar parts frequently assume critical proportions. Among the measures to avoid cracking we can recommend an increased pliability of moulding sand mixture, the proper choice of the temperature and pouring rate and proper care of the mould after pouring.

Steel foundry production can be attended to only by highly qualified personnel who have a thorough knowledge of the complex process of steel castings formation and know how to concentrate on the most important aspects.

The heat content and contraction of steel depend largely on its chemical composition. For the given grade of steel they remain constant, being influenced only by temperature variations. For this reason, temperature control can hardly be overestimated in the production of steel castings.

The temperature of the metal predetermines the amount of volumetric shrinkage, the degree of oxidation during pouring, fluidity, resistance to cracking and other properties of steel (see Chapter II).

The manufacture of pattern and moulding fixtures for steel castings is a study of itself (see Chapter III).

Moulding and core materials (more refractory and pliable) should be excavated, if possible, from nearby quarries. These materials must be such as to reduce to the minimum metal penetration and rat tails on the surface of the castings. The heat absorbing capacity of mould walls should be controlled by selecting proper moulding and core sand mixtures. Wood flour is sometimes added to the mixture to reduce heat conduction and increase the pliability of the mixture. Heat conductivity is sometimes promoted by using pig iron blasts. Numerous successful experiments have been conducted for this purpose in the U. S. S. R.

Separate parts of the casting are cooled by means of external chills and thick cast iron pipes with evaporatory water cooling.

The gating system should be so designed as to ensure adequate filling of the mould with metal and proper feeding and distribution of the liquid steel heat throughout the casting.

The arrangement of risering requires high qualifications of the production engineer and the moulder who design and make the mould. The metal expended on risering increases the size of the mould and labour input.

The risers are cut off after casting is completed, involving additional production costs. For this reason, steel castings come at about twice the price of pig castings of the same type and weight.

The heterogeneous structure of cast carbon or alloy steel which is the result of selective freezing of steel (solution of carbon in iron) can be eliminated by homogenising which increases the ductility and impact strength of steel. To increase its strength, the steel is normalised or hardened after annealing, and then tempered to improve its plasticity and ductility. Heat treatment of complex alloy steel is intended for stabilising, ageing or improving its structure.

Heat treatment requires specially built furnaces, electric power and greater labour input which also increases the costs of steel castings. Only unimportant castings, such as floor plates, weights for moulding boxes, etc., may be delivered untreated. Heat treated steel castings—large side plates of car bodies, tramway spiders and railway switches—are straightened with powerful hydraulic presses.

The external flaws in castings are dressed and welded thereby increasing the yield of good ingot and improving the technical and economic indices of production.

5. CARBON STEEL CASTINGS

Carbon essentially affects the structure, physical, chemical and mechanical properties of steel as well as its casting properties—fluidity, resistance to cracking, shrinkage and oxide-film formation. Many of these properties can be changed within a broad range by varying carbon content in steel. Besides, carbon is the cheapest alloying element, which determines the hardness or wear-resistance of steel but, when taken in increased amounts, it tends to reduce its ductility and impact strength.

An alloy with less than 2 per cent carbon is called steel and with more than 2 per cent—pig iron. In pig iron, either structurally free cementite or carbon, in the form of graphite, precipitates from the solution, or else austenite + graphite or austenite + cementite eutectics appear.

Exceptions are not infrequent, however. Thus, complex carbides may precipitate from the solution in alloy high-carbon steel. This is carbide or ledeburitic steel. An alloy containing 12-14 per cent Si, and only 0.2-0.6 per cent C, called ferrosilide, can be melted in the cupola and poured at 1300° C. It often contains a graphite ₐkish.

In its physical properties, liquid steel is less homogeneous than pig iron. It contains gases and nonmetallic inclusions and sometimes complex refractory carbides. Undissolved carbides deteriorate the continuity and homogeneity of a solid alloy and tend to reduce the ductility and strength of steel.

To obtain a uniform structure of liquid steel poured into foundry moulds, the temperature of low-carbon steel must be raised to 1630-1640° C before tapping and to 1580-1600° C in case of high-carbon steel. The metal should be thoroughly deoxidised in a furnace bath before tapping, or in the ladle, and poured at a proper rate. Vacuum built up in the furnace and the ladle as well as in the mould helps to free molten steel from gases and nonmetallic inclusions. This has become a regular practice during the recent years.

The solidification of low-carbon steel first yields crystals of refractory austenite and the δ-phase (peritectic transformation) and then most fusible intercrystalline componets—sulphides and phosphides, as well as surplus gases and nonmetallic inclusions. The more carbon there is in the residual liquid steel, the lower is the melting point and the higher the fluidity. In other words, dissolved carbon improves the casting properties of steel as do phosphorus and copper. Even in small quantities copper is capable of forming independent fusible phases and "grease" the austenite grains in the flow. Fusible phosphides have the same effect.

Although sulphides exhibit a high fusibility, they somewhat retard the flow of molten steel by force of their high viscosity in which they resemble a thickening tar. The fluidity of low-sulphur steel is quite satisfactory at a low superheat.

Due to a small specific gravity of sulphides and certain carbides, large-size castings exhibit micro- and macrosegregation which increases the heterogeneity and reduces the ductility of metal. For this reason, sulphur and phosphorus are classed among harmful impurities which are avoided by the choice of pure charge and by means of dephosphorisation and desulphurisation of the metal being melted, should the charge contain excessive amounts of sulphur or phosphorus, or by introducing certain alloying elements.

Carbon steel is stronger and more ductile than ordinary pig iron and most of nonferrous alloys but it is inferior in this respect to alloy steel. According to an appropriate Soviet standard, castings made of carbon steel are divided by quality into three groups (normal, improved and special) and by their chemical composition into nine groups.

The maximum content of sulphur and phosphorus in steel is determined depending on the quality group of castings and the melting method applied.

The content of residual admixtures (chromium, nickel and copper) should not exceed 0.3 per cent of each component.

Another Soviet standard establishes the mechanical properties of steel for as-normalised and as-annealed castings depending on the grade of steel. After hardening and tempering, the mechanical properties are established by a mutual agreement between the user and the manufacturer.

In the Federal Republic of Germany, the standard DIN 17006 prohibits a sulphur or phosphorus content in excess of 0.04 per cent of each, irrespective of the type of steel smelting equipment. No more than 0.35 per cent carbon should be present in cast-weld constructions.

Test samples for chemical composition control are taken during pouring in conformity with an appropriate standard. Two test samples are taken from each ladle after half of the metal has been poured from the ladle. The test samples are poured into cup-shaped moulds 120 mm high, 50 mm in diameter at the bottom and 80 mm at the top, left unfilled by 20 mm. Some aluminium is added to the test sample, which weighs not less than 2 kg, and a tag bearing the number of heat, the number of test and the stamp of the technical inspection department is placed into the cup-shaped mould.

Arbitration chemical analysis is done in conformity with another State standard.

Test bars for testing mechanical properties are cast in the middle of pouring both in dry sand and green sand moulds. If castings are made in metal moulds the bars are also cast in metal moulds. The test pieces can be cut from the casting at the section about the same thickness as the test bar. The results of such test hold for all the castings of the given lot. On agreement with the user a test piece may be cast integral with the casting.

Tensile tests performed on pieces 10 mm in diameter with a design length of 50 mm and impact strength tests also conform to appropriate State standards. The mechanical properties of any alloy, including steel, depend on macro- and microstructure as well as on contraction, fluidity, hardenability and resistance to cracking. All these properties must be made the subject of close study to attain high strength and ductility of steel and eliminate the flaws.

6. CASTINGS OF ALLOY STRUCTURAL STEEL

. The effect of each alloying element on the properties of steel is studied in detail in the course of physical metallurgy dealing with the structure and properties of rolled and forged articles. The course "Foundry Practice" deals with the formation of a cast structure not subjected to rolling or other types of heat treatment, i. e., it studies the effect of alloying elements on the engineering properties, including the primary solidification of steel.

Let us consider in general the effect of alloying elements on steel while it passes from a liquid into a solid state on the basis of the general theory of alloying.

The most important factor is here the reaction of an alloying element to iron, i. e., the ability of this element to combine with iron and the admixtures that go with it to form chemical compounds or limited and unlimited solid solutions. This ratio greatly affects the degree of the physical homogeneity of a liquid and solid alloy.

The course of physical metallurgy tells us that the solubility is determined by the size of the atoms of the substance being dissolved and the solvent. Nickel, chromium, cobalt and vanadium form unlimited solutions since the difference in their atomic radii does not exceed 14 per cent (diameter of α-iron atoms is 2.56 and of γ-iron 2.556 Å). Silicon, manganese, tungsten, molybdenum, etc., combine to yield limited solutions.

With the difference between the atomic radii not above 8-9 per cent the mutual solubility of the elements in a solid state can be boundless indeed. The solubility will be restricted with the difference more than 8 per cent, but not above 15-16 per cent. If the difference is in excess of 16 per cent, solid solutions will either not form at all or the region of their propagation will be very small.

The elements forming free phases during the solidification of steel, for example carbides, carbonitrides and oxides, reduce the physical homogeneity and increase the ductility of solid alloys. Free phases may be the cause of excessive brittleness. Oxides form in a liquid alloy and tend to decrease fluidity, especially the oxide films of titanium and aluminium.

The structurally free phases of undissolved elements form segregations which are liable to reduce not only the ductility but the strength as well.

The strength of alloys increases in inverse proportion to ductility.

The alloys progressively gain in strength, the more distorted is the crystal lattice due to the presence of various dissolved elements.

The maximum strength is attained from complex alloying, i. e., by adding several alloying elements, with the percentage by weight of one admixture increasing the strength of the alloys far less than the identical percentage amount of several admixtures. This accounts for the predominant use of multicomponent alloy steels, both structural and heat-resistant, such as grades 27ХГСНМЛ, 35ХГСНМЛ, etc.

Let us consider the general rules for the selection of alloying elements for steel to be cast.

1. Increased strength and ductility of steel to reduce the weight of the whole assembly, prolong its life and improve the productiv-

ity, speed, safety and reliability in operation as, for example, when designing jet engines, high-speed electric and diesel locomotives, electrical equipment, atomic engines, etc.

2. Inexpensive and readily available alloying elements due to favourable conditions of ore extraction, concentration and other economic factors of iron and steel production. These features place the alloying elements into the following order: manganese, silicon, chromium, aluminium, nickel, tungsten, titanium, vanadium, molybdenum, cobalt, zirconium, niobium, cerium and lantanum.

When combined with iron, manganese forms unlimited solid solutions. It also combines with carbon to produce manganese carbide, Mn_3C, which with cementite Fe_3C will yield complex carbides $mMn_3C \cdot Fe_3C$.

Of the several manganese oxides—MnO, Mn_3O_4 and MnO_2—MnO is the most important in steel smelting. The affinity of manganese for oxygen at ordinary pressures is more pronounced than in iron. Manganese tends to deoxidise iron, producing MnO. This oxide is distributed between the metal and slag. In metal, it yields weak limited solutions and in slag—unlimited. Pure manganous oxide is neutralised by acid oxides, for example silica, forming colloidal particles of $mMn \cdot nSiO_2$. Silicates of manganese tend to coagulate in a liquid metal and form larger particles which are removed from the metal the faster, the larger is their diameter. Tephroite $2MnO \cdot SiO_2$ melts at 1320° C and forms in steel liquid droplets readily rising to the surface.

Manganese forms an important sulphide MnS which neutralises the harmful effect of the fusible iron sulphide FeS.

These chemical properties of manganese determine its usage as a deoxidiser and an alloying element. The ordinary carbon steels vary in manganese content between 0.35 and 0.8 per cent. In structural alloy steels manganese is present either as an independent alloying element: 1.1 to 1.6 per cent (grades 27ГЛ and 35ГЛ), together with silicon (grades 20ГСЛ and 30ГСЛ), or in combinations with other alloying elements (grades 30ГСТЛ and 35ХГСЛ, etc.).

The low hardenability of manganese steel limits its application to castings with a wall thickness not in excess of 100 mm.

In α-iron manganese (up to 0.10 per cent) forms a substitutional solid solution. A greater manganese content in the solution adds to the ferrite hardness (65-140 *Bhn*).

A higher manganese content in steel improves its strength. The relative elongation of steel reaches its maximum with 2 per cent Mn, while the impact strength declines.

Table 1 illustrates the mechanical properties of manganese steel used for couplers and greased gears operating in friction. This steel is gaining ever wider applications.

Table 1

Mechanical Properties of Cast Manganese Structural Steel

Grade	Chemical composition, per cent			Mechanical properties					
	C	Mn	Si	σ_s, kg/mm²	σ_b, kg/mm²	δ_s, %	ψ, %	a_k, kgm/cm²	*Bhn*
25 ГЛ	0.2-0.3	1.1-1.9	0.3-0.45	30-45	50-70	30-50	45-50	10-12	160-270
30 ГЛ	0.25-0.35	1.05-0.35	0.3-0.45	30-40	50-60	28-55	45-50	7	170
40 ГЛ	0.35-0.45	1.35-1.55	0.3-0.45	35-40	55-65	20-25	40-50	6	200
50 ГЛ	0.45-0.55	1.40-1.60	0.3-0.4	40-50	60-75	15-20	35-45	4	240

The use of manganese steel instead of carbon steel makes it possible to increase the weight of a train by force of a much higher tensile strength of the coupler heads of railway rolling stock. Instead of 170 tons needed to rupture carbon steel coupler heads, the experimental heads made from structural alloy steel of grade 25 ГЛ, broke only under a force of 400-450 tons.

Silicon combines with iron to form silicides which prevents atmospheric oxidation, thereby improving the corrosion resistance of castings. With oxygen silicon yields the monosilicide SiO, which converts to a stronger SiO_2. The affinity of silicon for oxygen is greater than in iron and, consequently, silicon behaves in steel as deoxidiser. During the process of deoxidation it forms harmful refractory oxides which enter the steel in the form of nonmetallic inclusions.

Under certain conditions silicon may force carbon out from the solution. The steel develops black fractures and for this reason silicon cannot be used by itself as an alloying element. To make structural steels silicon is combined with other alloying elements— manganese (grade ГС), and chromium with manganese (grades ХГС, ДХСН).

Copper and iron form limited solutions, which promotes the anticorrosive resistance of steel. Alloying of steel with copper is

* σ_b = tensile strength;
σ_s = yield point;
δ = elongation;
ψ = reduction of area;
a_k = impact strength;
Bhn = Brinell hardness.

made a wide use of in heavy engineering to produce large castings, but copper steels are rather limited in their applications.

At the same time, copper, like manganese, can be employed as a good substitute for nickel but in more complex multicomponent steels, such as grades 30ДХСНЛ and 150Д2СХЛ.

Like manganese, silicon and copper, chromium is relatively cheap and has a fairly widespread use as an addition. A great variety of structural steels are alloyed with chromium alone or together with other elements.

Silicon and chromium tend to narrow down the region of solid solutions in γ-iron at 19 per cent Cr and 0.6 per cent C, the γ-region disappears and the steel becomes ferritic. By varying carbon content in chromium steel we can obtain steel with phase transformations, i. e., steel of the pearlitic class, and without these transformations—austenitic or ferritic steels. When the steel is very rich in carbon it forms complex chromium carbides and becomes either carbidic or ledeburitic. Structural chromium steels belong to the hypoeutectic class.

Chromium sharply increases hardenability and shifts the eutectoid point S to the left. At a slow critical rate of hardening, large-size castings with uniform mechanical properties can be obtained from chromium steel. This steel is as strong as carbon steel while its ductility is much greater. The evolution of dispersed carbides on tempering makes chromium steel highly resistant to wear. In addition to chromium, more complex structural steels contain nickel, copper, molybdenum or silicon and manganese, or nickel and tungsten, nickel and titanium, and are remarkable for their high strength and ductility.

Nickel combines with iron to form a continuous series of solutions. When present in a solution of α-iron, nickel distorts the crystal structure and increases the strength of the alloy involving a comparatively small reduction of ductility. It tends to increase supercooling and slows down the critical rate of hardening. With 5 per cent Ni, the steel assumes a martensitic structure when subjected to hardening in a jet of compressed air. An increase in the hardenability of steel produces a homogeneous structure in heavy sections and promotes the advantages accruing from normalising, used instead of hardening, in a liquid medium with a subsequent tempering.

Like silicon, nickel ousts carbon from the solution. Prolonged tempering may cause free carbon to evolve, thereby favouring the formation of black fracture. This shortcoming restricts the use of nickel steels for castings. Although nickel is rather expensive, it is extensively used in combinations with other alloying elements such as chromium, tungsten, molybdenum and manganese to make important castings because it greatly improves the strength and hardenability of steel.

Chromium-nickel steel had been used by foundrymen as far back as the beginning of our century. Today, its use has been drastically reduced, however, due to the appearance of other still more complex structural steels. The mechanical properties of steel in castings 25 to 200 mm thick are nearly the same. Temper brittleness is a major shortcoming of this steel. It will be well advised therefore to harden the castings after preliminary annealing in a medium heated to the tempering temperature (300-500° C). This process is called austempering.

Chromium-nickel steel contains a broad variety of admixtures which widens the applications of castings to be case-hardened at a low carbon content (0.10-0.15 per cent) or castings which are to operate in dry friction (0.9-1.1 per cent C). The presence of manganese in steel is useful at 0.60 per cent and upward, of silicon from 0.3 to 0.7 per cent, of chromium from 0.5 to 2 per cent and of nickel from 1 to 1.9 per cent. After normalising and tempering from 580 to 650° C the mechanical properties of castings made of steel 30ХНЛ to 40ХНЛ vary within the following range: σ_s =30-40 kg/sq mm, σ_b=50-70 kg/sq mm, δ_5=10-16 per cent, ψ=20-25 per cent, Bhn=140-269, a_k=3.5-6.0 kgm/sq cm.

This steel goes to make shoes in machines, parts and links of chains, parts for oil refinery equipment, rolls, parts for diesel and electric locomotives which should be very strong, ductile and light. The steel of this type belongs to the second strength category.

Steels 35ХНЛ and 40ХНЛ are used to cast toothed rims, bushes and gears for earth diggers and horizontal rolls for slab mills.

Molybdenum is much more expensive than nickel. But even such small additions of molybdenum as tenth fractions of a per cent will noticeably increase the strength, ductility and hardenability of steel and will make it much more heat- and wear-resistant and corrosion-proof. Molybdenum goes mainly for the manufacture of heat-resistant steel because it tends to stabilise its structure and slows down the ageing. Molybdenum steel with only 0.3 per cent Mo and 0.05-0.10 per cent V is in the second strength category and its impact strength does not exceed 10 kgm/sq cm.

Tungsten costs less than molybdenum and is a very good substitute for it. It finds an ever greater application to make cast tools mainly by the investment pattern method and heat-resistant castings.

Vanadium refines the primary steel grains, reduces the temper brittleness and increases the strength and ductility of steel. Formerly obtained from blast-furnace slags, vanadium is now steadily growing lower in cost and its field of application is progressively increasing.

More and more titanium is being produced with each passing year and its cost is gradually brought down, like that of vanadium.

It binds nitrogen dissolved in steel in the form of titanium nitrides but, due to a greater affinity for oxygen, it tends to develop oxide films. Its substitutes are niobium, cerium and some other elements.

Aluminium deoxidises the steel and brings into it nonmetallic inclusions, but it is still used as an alloying element tending to increase the resistance to scale formation in steel castings. Structural aluminium steels are rather rare.

Below are several examples of the composition and mechanical properties of a multicomponent structural steel. Its most conspicuous feature is the increase in strength, when it is alloyed with several components added in small amounts.

Apart from the usual strength and ductility, some grades of multicomponent steel exhibit an increased wear resistance as, for example, steel 30ХГСЛ. According to Y. Nekhendzi, this steel consisting of 0.3 per cent C, 1.2 per cent Mn, 0.7 per cent Si and 0.7 per cent Cr has the following properties after being normalised from 880-900° C and tempered by 400-430° C: $\sigma_s \geqslant 35$ kg/sq mm, $\sigma_b \geqslant 63$ kg/sq mm, $\delta_s \geqslant 14$ per cent, and $\psi \geqslant 30$ per cent.

The attempt to obtain steel of the third category of strength, without nickel and molybdenum, prompted the investigators (K. Grechin, N. Tuchkevich et al) to develop the Л35ХГСА Chromansil grade for important castings. This steel contains 0.32-0.40 per cent C, 0.6-0.9 per cent Si, 1.0-1.2 per cent Mn, 0.7-1.0 per cent Cr; $Ni \geqslant 0.40$ per cent, $S \geqslant 0.30$ per cent, $P \geqslant 0.035$ per cent, $V \geqslant 0.1$-0.2 per cent.

For a number of years now the castings made from aluminium and magnesium alloys are being superseded at Soviet plants by more reliable castings made from structural multicomponent, high strength steels. The steel is smelted in induction arc furnaces. Two Soviet-made grades of steel proved up to the mark— Л35ХГСА and Л27ХГСНА. The chemical composition of the former grade varies within 0.32-0.40 per cent C, 0.6-0.9 per cent Si, 1.0-1.2 per cent Mn, 0.7-1.0 per cent Cr; $N \geqslant 0.40$ per cent, $S \geqslant 0.030$ per cent, $P \geqslant 0.035$ per cent, and 0.1-0.2 per cent V.

Prior to rough machining, the castings should be annealed or normalised at $920 \pm 20°$ C, holding them for two hours.

The mechanical properties of this steel as found from the heat-treated test pieces cut out from test bars should conform to the following requirements:

(a) after hardening from $890 \pm 10°$ C in oil and tempering at 630-670° C: $\sigma_s \leqslant 60$ kg/sq mm, $\sigma_b \leqslant 80$ kg/sq mm, $\delta_5 \leqslant 12$ per cent, $\psi \leqslant 30$ per cent, $a_k \leqslant 4$ kg/cm², $Bhn = 4.05$-3.75 mm;

(b) after hardening from $890 \pm 10°$ C in oil and tempering at 570-630° C: $\sigma_s \leqslant 85$ kg/mm², $\sigma_b \leqslant 100$ kg/mm², $\delta_s \leqslant 8$ per cent, $\psi \leqslant 20$ per cent, $a_k \leqslant 2.5$ kgm/cm², $Bhn = 3.6$-3.35 mm.

Grade Л35ХГСА steel represents the third category of strength.

The other grade Л27ХСНА with nickel belongs to the fourth or fifth category of strength. The chemical composition of this steel should lie within the following limits: 0.24-0.30 per cent C, 0.5-0.8 per cent Si, 0.9-1.2 per cent Mn, 0.7-1.0 per cent Cr, 1.4-1.8 per cent Ni; Si$\geqslant$0.035 per cent, and P$\geqslant$0.035 per cent.

The mechanical properties of the steel are:

(a) after hardening from 890$\pm$10 °C and tempering at 200-240° C: $\sigma_s \leqslant$120 kg/mm^2, $\sigma_b \leqslant$150 kg/mm^2, $\delta_5 \leqslant$6 per cent, $\psi \leqslant$20 per cent, $a_k \leqslant$2.5 kg/mm^2, Bhn=30-2.70 mm;

(b) after hardening from 890$\pm$10° C and tempering at 280-320° C: $\sigma_s \leqslant$100 kg/mm^2, $\sigma_b \leqslant$130 kg/mm^2, $\delta_5 \leqslant$7 per cent, $\psi \leqslant$25 per cent, $a_k \leqslant$2.5 kgm/cm^2, and Bhn=3.2-2.85 mm.

The chemical analysis is performed in conformity with an appropriate State standard and the sample is taken during pouring of steel. Each heat should be sampled at least for two test bars. Tensile tests are performed on pieces of any diameter with the design length of the l: d=5. Impact strength is also checked by an appropriate State standard.

CHAPTER II

CASTING AND OTHER ENGINEERING PROPERTIES OF STEEL

A sound and effective design of machine elements depends on the production methods employed in rough stock departments of engineering plants and on the engineering properties of the materials from which they are made. Among the basic engineering properties of steel are machinability of castings by sharp cutting tools, weldability, hardenability and a number of others. Casting properties distinguish liquid steels during the casting process. These are fluidity, atmospheric oxidation, volumetric and linear shrinkage, formation of oxide films, shrinkage cavities, porosity and freedom from cracking.

The casting properties depend on the chemical composition of an alloy and such physical constants as surface tension, ductility, density, thermal and electric conductivity as well as saturation with gas and the freedom of the molten metal from nonmetallic impurities during pouring into moulds.

The casting properties vary within a very narrow range which can be considerably broadened at will by changing the temperature and pressure.

7. FLUIDITY OF STEEL AND MOULD FILLING ABILITY

One of the prime tasks in foundry production is to make a liquid alloy heated to a definite temperature so fill the mould that its configuration corresponds exactly to that of the pattern.

The physical properties of liquids (transparency, specific gravity, viscosity, wettability, surface tension) cannot alone satisfy all the requirements made to alloys in casting where temperature or pressure undergo a rapid change. A knowledge of the so-called engineering properties of alloys, such as fluidity, for example, is required.

A. Portevin and P. Bastien (France) were the first to give a classic interpretation of fluidity and to express it in mathematical categories:

$$\lambda = \alpha \frac{cd(\theta_f - \theta_s)}{\theta_f - \theta_m} + \beta \frac{dL}{\theta_f - \theta_m} \; mm, \qquad (1)$$

where λ = fluidity, mm;

α = coefficient depending mainly on the properties of an alloy, including its viscosity;

β = coefficient depending mainly on the properties of the mould, including its heat conductivity, wettability and roughness;

c = heat capacity of molten alloy, cal/cm³ °C;

d = density, g/cm³;

θ_f = temperature during the filling, °C;

θ_s = solidus temperature, °C;

θ_m = temperature of the mould, °C;

L = part of the latent heat of solidification, cal.

This equation has the disadvantage of the coefficients α and β having too broad a physical meaning found empirically and of the lack of their connection with viscosity and surface tension.

Neither are they related in any way with hydrodynamic properties of liquids and channels along which the liquids flow. On the other hand, the formula is valuable in that it establishes the relationship between fluidity and the temperatures of the alloy and the mould in pouring, but the experience goes to show that fluidity is a function of the first power of the alloy superheat at a comparatively small superheat of 100-150° C above the solidus temperature. At a higher superheat the fluidity practically does not increase and tends asymptotically towards a constant value.

Many scientists made steel fluidity their subject of study. There was a great variety of mould designs for fluidity tests and each new research in this field revealed new dependences of steel fluidity on new, formerly unknown, factors. Thus, S. Kantenik of the U.S.S.R. used a U-shaped test piece to determine the fluidity of steel as a function of the size, number and shape of dispersed nonmetallic inclusions. The fluidity increased, as more and more inclusions, especially sharp-edged, were removed from the steel (Fig. 1).

The Curry fluidity spiral in Fig. 2 for tests at a constant pressure head and filling rate has gained the widest application for measuring steel fluidity. To make the process more effective, the test piece was made of the same material as the foundry mould and installed with the aid of a water level. W. Ruff suggested a straight-bar type test piece of round cross section (Fig. 3). The helix developed by V. Gruzin and V. Saveiko (Fig. 4) is employed for steel fluidity tests in a furnace bath. The two halves of the helix are brought together and immersed into the molten metal after deslagging.

These fluidity tests give an idea of the ability of the metal to flow through spiral, straight or helical passages of a round or nearly round cross section, depending on the temperature of the metal in pouring, the temperature of the mould at the start of pouring, and, partly, depending on the nature of the alloy solidification.

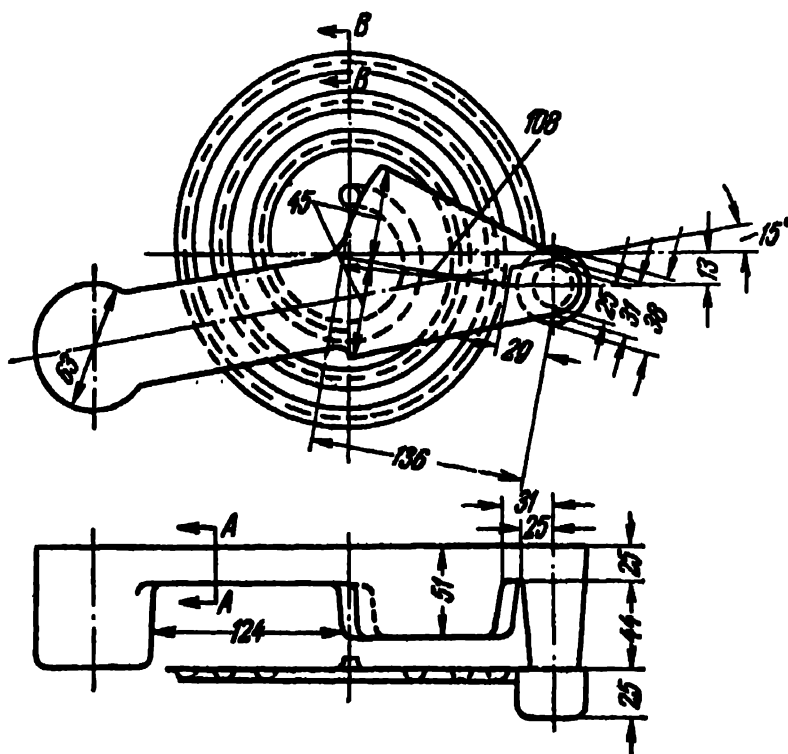**Fig. 1.** U-shaped piece for fluidity test (Nekhendzi, Samarin, Kantenik)
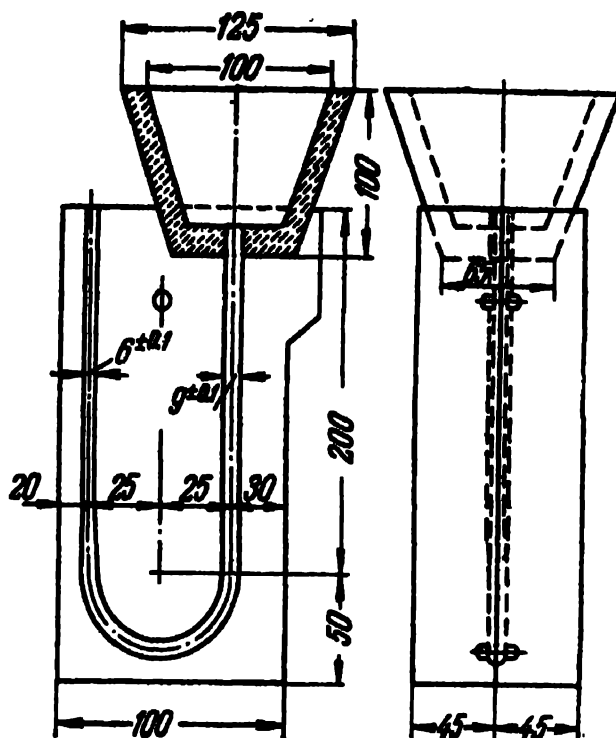
Fig. 2. Fluidify spiral used for tests at a constant pressure head and filling rate (Taylor, Rominski, Briggs)

A. Bochvar has found that the best fluidity is attained in alloys which form equiaxial crystals and the poorest in alloys which form dendritic crystals. The latter include alloys, steel among them, which solidify within the temperature range. Steel is lower in fluidity than pig iron which possesses a certain amount of eutectics. Bochvar believes that the metal stops flowing when a continuous skeleton of solid-phase crystals is formed in the flow being cooled. This state is known as zero fluidity. Besides, from the viewpoint of statics, the metal stops flowing when the strength at the head of the flow becomes equal to the pressure head. Hence the conclusion that fluidity depends not only on the above factors but also on the thermal parameters of the alloy and the hydrodynamic properties of the test piece or the foundry mould
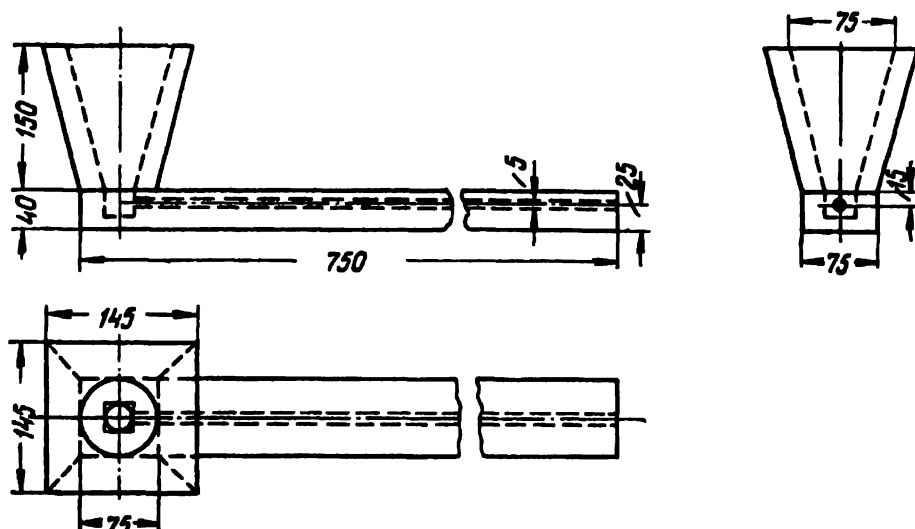
Fig. 3. Straight-bar type test piece of round cross section (Ruff).

The rounded head of the solidified metal (Fig. 5) shows that the remains of the liquid phase came through a sieve of hard crystals and assumed the form of the last drop which had sunk to the bottom of the test piece.

Fluidity tests can be conveniently used to determine whether the alloy is ready to be tapped into the ladle in order to avoid short-run castings due to excessively cold metal.

The metals (aluminium and steel) that readily oxidise in the air may possess insufficient fluidity even in a superheated state. Timely tests can prevent spoilage and production losses. Films and slag which found their way into the test piece together with the metal considerably reduce fluidity. So far we have a very scarce knowledge of the effect produced by viscosity and surface tension on fluidity we cannot disregard these factors.

The advantage of fluidity tests lies in the fact that they can be used to find final results of the effect both the known and unknown factors exert on the metal taken for the, test. Some condi-

tions can be determined only on the basis of a profound differential analysis of the occurring phenomena.

Comparison of the fluidity tests of steel poured in at the same superheat above the solidus temperature shows that fluidity increases
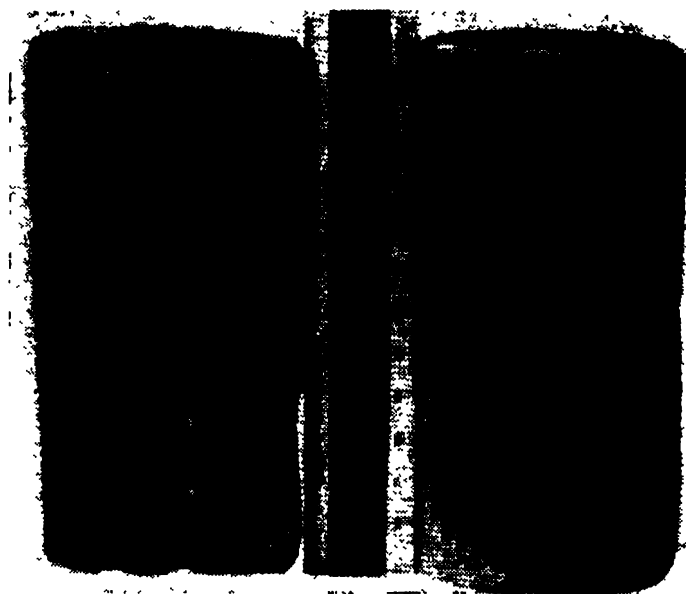
Fig. 4. Fluidity test helix (Gruzin, Saveiko)

with a smaller solidification · interval. Steel of a peritectic composition with 0.20 per cent C has the lowest fluidity. When increased to 1 per cent, silicon tends to reduce fluidity, and then
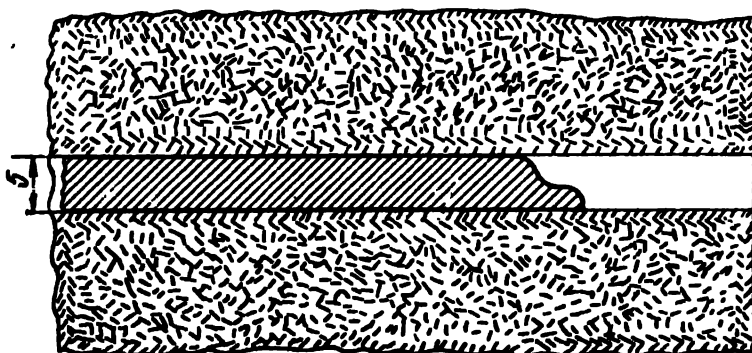
Fig. 5. Head of the metal solidified in the test piece (diagram)

increases it as its content rises. Manganese, copper, nickel, niobium and phosphorus increase fluidity, while aluminium, chromium, titanium and sulphur reduce it. When the temperature is raised to 1600° C the fluidity of certain grades of steel goes up and then its increase is gradually retarded.

The above test pieces with round or nearly round passages provide no comparable results when thin-walled and large-size castings are made, even when the test piece and the mould are of the same material. This can be attributed to the different hydrostatic and hydrodynamic conditions which obtain when round and flat-section moulds are filled.

The Reynolds equation shows that the liquid in a flat duct passes faster from the turbulent motion to a steady motion than in a round pipe:

$$Re = \frac{v\varrho}{\nu},\tag{2}$$

where Re = Reynolds number;
v = linear velocity of motion, cm/sec;
ϱ = hydraulic radius of the pipe equal to 1/4 of the diameter for a round pipe, cm;
ν = kinematic viscosity of liquid, m²/sec.

Solving Eq. (2) with respect to v we obtain

$$v = \frac{Re\,\nu}{\varrho}.$$

The velocity v is the length of path λ divided by pouring time τ whence

$$\lambda = \frac{Re\,\nu}{\varrho}\tau.\tag{3}$$

A hot liquid in a laminar flow gives up heat faster than in a turbulent flow and therefore the viscosity factor in the numerator of Eq.(3) will quickly decrease and reduce the value of λ. A liquid flowing in a flat channel will stop sooner than in a round pipe, given the same cross sections. These considerations require that the metal be fed to large-size castings via round, and not flat, channels.

Fluidity is also affected by pressure head since

$$\lambda = v\tau = \mu\tau \sqrt{2gH},\tag{4}$$

where μ = coefficient of hydraulic resistance;
g = gravity acceleration, cm/sec²;
H = pressure head, cm.

To obtain comparable results from fluidity tests the test piece should be poured with the pouring cup filled to the brim.

When comparing the fluidity of test pieces made in sand and metal moulds with an alloy of the same composition and at the same pouring temperature we shall not fail to see that the fluidity in the metal mould will be lower than in the sand one. In other words, a new factor comes into play which depends on the properties of the mould and not of the metal.

The fluidity of the metal poured into a cold mould for testing wil! be lower than if the metal is cast into a hot mould. If the test mould is heated to the pouring temperature, then the denominator of both terms of Eq. (1) will become zero and the fluidity—infinity. Hence the practical conclusion that to obtain thin-walled castings the moulds should be heated prior to filling.

The effect exerted on fluidity by the mould makes it necessary to supplement the notion of fluidity with another, no less arbitrary, notion of mould filling ability. Then the first term in Eq. (1) will express the ability of the metal to flow (fluidity) and the second term—the ability to fill the mould. In the light of this new information, both properties will differ from each other despite their close interdependence.

A. Spassky suggested what he called a ball test (Fig. 6) to measure the fluidity from the size of the unfilled hole under a ball placed on a surface. J. Czikel (Germany) developed his "harp" test piece (Fig. 7) with the help of which fluidity could be determined much more effectively than by spiral and other fluidity tests. In this design thick and thin vertical channels are arranged at equal distances from the sprue. The length of the channels feeding metal from the sprue and the angles of turn of the gates are also the same and, for this reason, the channels of various diameters are in similar conditions from a hydraulic point of view The mould filling ability is indicated by the angle of inclination of the line passing over the tops of the cast rods when the test piece is presented as developed on a flat vertical plane.

J. Czikel's test recreates better the analogous hydraulic and thermal conditions of the metal flow through the channels inside the test piece, with identical hydraulic resistances along the length of the horizontal channel for thin and large channels. A step forward in establishing the mould filling ability of liquid alloys was made by Wood and Gregg of England who determined the difference between the fluidity and mould filling ability used for high-nickel and other nickel-containing alloys. In their experiments they used the fluidity spiral of H. Taylor, E. Rominski and C. Briggs (see Fig. 2). They measured the mould filling ability in per cent of the spiral length in that part of the test piece which had distinctly sharp corners. When the test piece was removed from a sand mould a part of the spiral beginning from the leading end had rounded corners and the rear end, adjoining the sprue, had sharp corners at the top (Fig. 8). The length of the spiral with sharp corners as taken in per cent to the length of the entire spiral is an arbitrary magnitude characterising the ability of a given alloy to fill the mould.

To be able to compare the results of determining fluidity and mould filling ability, the composition of the moulding sand
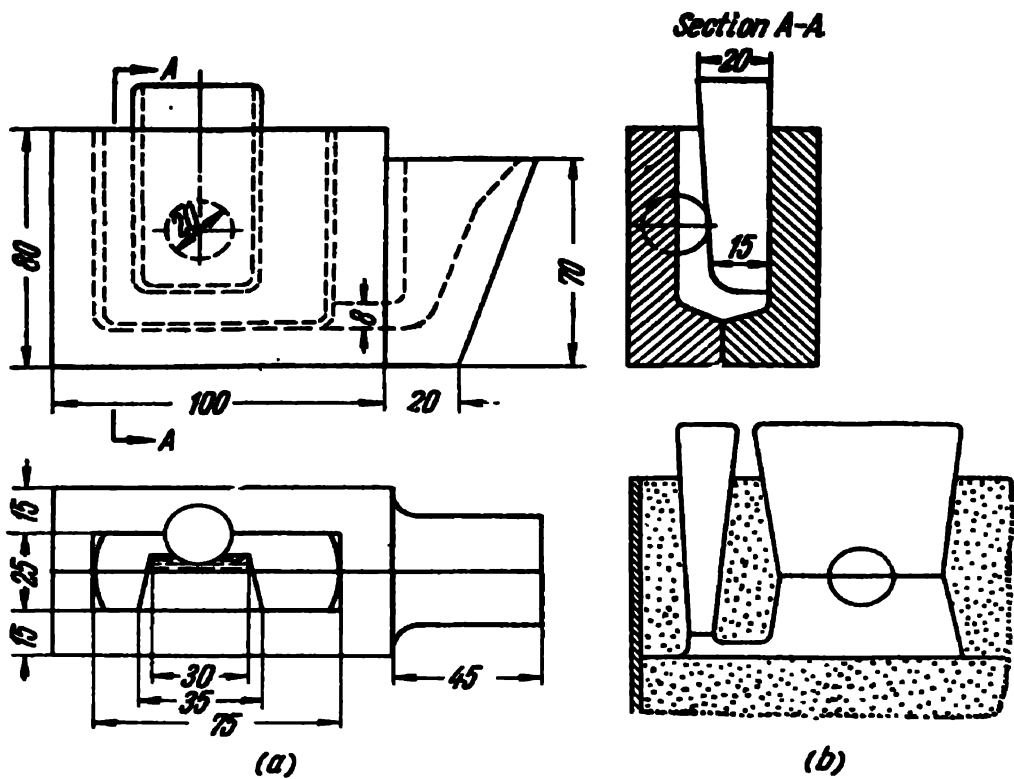
Fig. 6. A ball for fluidity tests (Spassky):
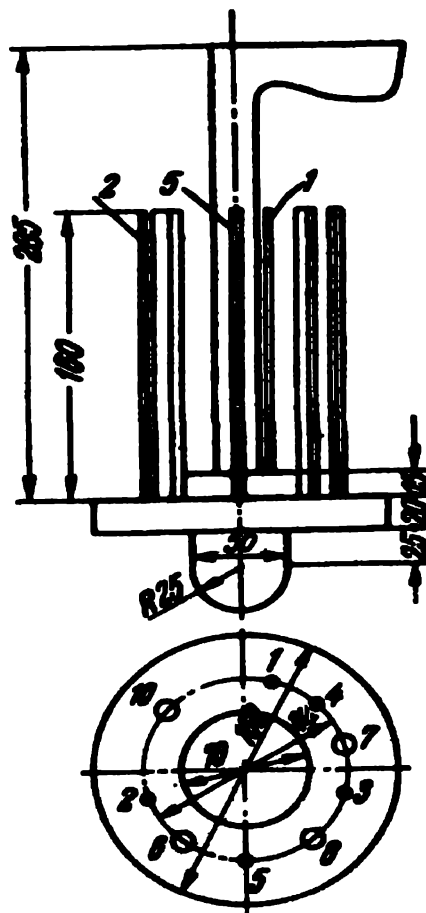*a*—metal mould; *b*—sand mould

Fig. 7. A "harp" for fluidity tests (Czikel)

mixtures used for the preparation of foundry and test moulds should be strictly identical.

Wood and Gregg experimented with the fluidity of high-nickel alloys whose composition is illustrated in Table 41. They also discovered how nickel-base alloys filled the moulds. The diagram in Fig. 9 shows that the fluidity of the Monel alloy (0.1 per cent C, 1 per cent Si, 1 per cent Mn, 30 per cent Cu, the rest Ni) is almost identical with the Nimonic 90 alloy (0.1 per cent C, 0.5 per cent Si, 0.5 per cent Mn, 1.2 per cent Al, 2.4 per cent Ti, 16 per cent Co, 20 per cent Cr, the rest nickel). Grade 18-8 stainless steel (0.05 per cent C, 0.8 per cent Si, 0.5 per cent Mn, 18 per cent Cr, 8 per cent Ni, the rest iron) is inferior to the two above alloys as regards fluidity and mould filling ability.

(a) *(b)*
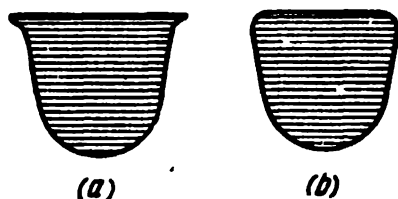
Fig. 8. Cross section of a
spiral for fluidity test:
a—at the beginning of spiral;
b—at the leading end of spiral

To obtain sound castings, grade 18-8 steel requires superheating in pouring by at least 150-200°C to prevent the formation of oxide films, which deteriorate fluidity and mould filling ability, and obtain homogeneous surface structure of steel with a highly corrosion-resistant skin.

When silicon content was increased from 0.5 to 2.5-3 per cent nickel-base alloys—Monel, Corronel B (0.05 per cent C, 0.5 per cent Si, 0.5 per cent Mn, 6 per cent Fe, 28 per cent Mo, the rest Ni), Nimonic 90, Nimonic DS (0.1 per cent C, 2 per cent Si, 1 per cent Mn, 18 per cent Cr, 27 per cent Ni, the rest Fe) showed a much greater mould filling ability even at a low superheat. This prompted Wood and Gregg to draw the conclusion that nickel-base and nickel-bearing alloys should be superheated to improve their fluidity. For the improvement of mould filling ability however, the change in the chemical composition of the alloy by increasing silicon content proved much more effective than superheating during pouring.

Even at a comparatively high superheat of steel with good fluidity the moulds for thin-walled and large-size castings may be filled inadequately due to short run. This can be attributed to phenomena which depend only on the properties of the mould, for example, on the change in the composition of the wash or the facing of the mould. To establish the full identity between the test piece

Fig. 9. Change in the mould filling ability(black dots) and fluidity (white dots) as a function of temperature:

1 and *5*—Monel metal; *2* and *4*—Nimonic 90; *3* and *6*— grade 18-8 stainless teel

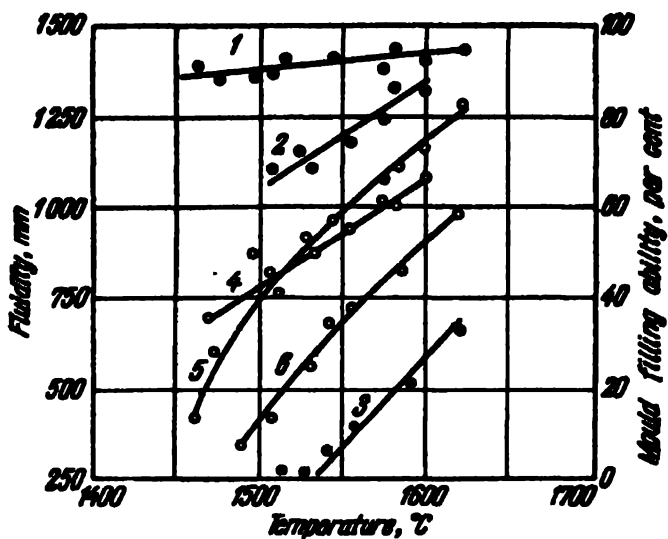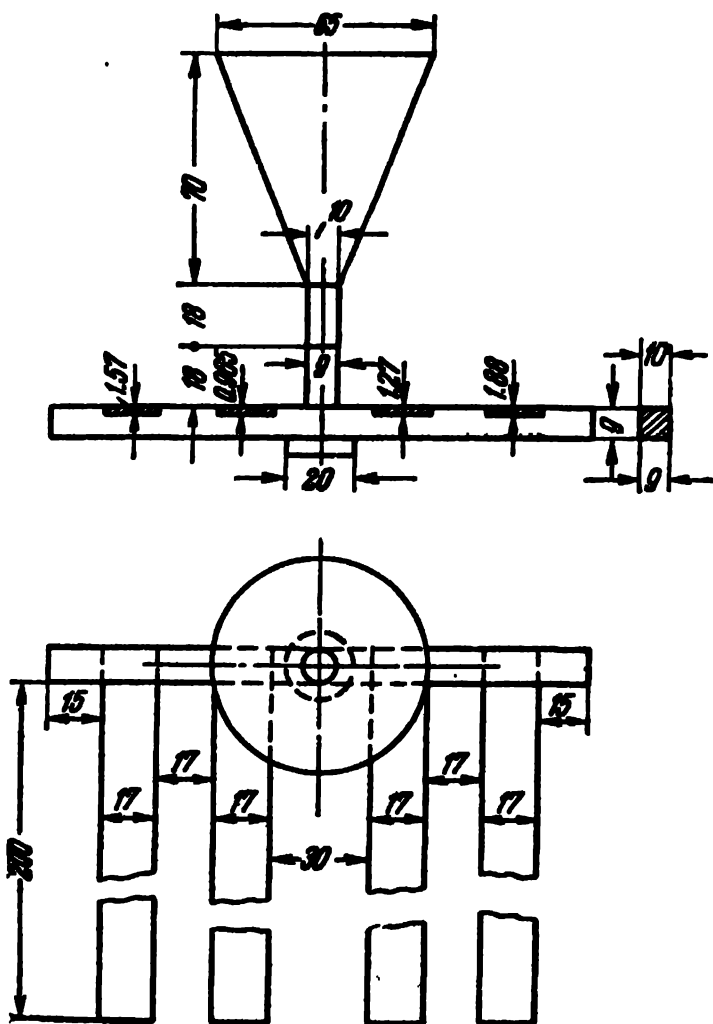*Fig. 10.* Kondic's strip-type fluidity and mould filling ability tests

and the mould for a thin-walled castings, V. Kondic* (England) suggested a strip-type test piece which could measure both fluidity and mould filling ability (Fig. 10) and in which the phenomena of wettability and surface tension are more prominent than in test pieces with round channels. The strip-type test pieces can be recommended for application at plants which produce thin-walled castings as in the aircraft industry, for example.

It should be noted in conclusion that some alloys and steels having very poor fluidity cannot be cast. Before introducing new alloys in foundry shops it is necessary to study them thoroughly and "adjust" their casting properties by changing somewhat the composition or structure of as-cast metal. Since many years of research may be required to complete such preliminary work, branch research institutes will be able to cope with this task much better than individual plants.

8. SATURATION OF STEEL WITH GASES. GAS CAVITIES AND FLAKES IN CASTINGS

Liquid iron and its alloys possess a high chemical activity and interact with furnace gases, slag and the linings of the furnace, spout, ladle and gating of the foundry mould. Chemical alterations occurring in molten metal in a furnace bath or during pouring promote the formation of gases which may either escape from the metal as bubbles, as for instance during oxidation, or be trapped in the metal.

The greatest effect on the casting and operating properties of steel is produced by oxygen, hydrogen and nitrogen.

Oxygen dissolves in steel as a chemical compound and does not obey the Sieverts Law for biatomic gases:

$$K_s = \frac{[G]}{\sqrt{P_{g_2}}}, \qquad (5)$$

where K_s = solubility constant, cm³/g;
  $[G]$ = gas concentration in the metal, cm³/g;
  P_{g_2} = partial pressure of gas above metal, g/cm².
According to the Gibbs Law,

$$F = C - P + 2,$$

where F = number of degrees of freedom;
  C = number of components;
  P = number of phases;

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* V. Kondic. A New Test for Fluidity, *Foundry*, V. 87, No. 12, 1959, p. 79-83.

the concentration of gas in metal at a constant temperature is a function of pressure. If the gas forms a chemical compound with the metal, a third phase appears, the system becomes monovariant and the concentration of gas in the metal at a constant temperature no longer depends on pressure. This shows that the metal has become saturated and that the compound begins to come out of the solution.

When the temperature reaches the melting point of iron (1535° C) the solubility of oxygen amounts to 0.21 per cent, or to 0.94 per cent in terms of ferrous oxide FeO; at 1700° C the solubility rises to about 2 per cent while at a temperature of 715° C only 0.11 per cent of oxygen is dissolved (0.49 per cent FeO). In an oxidised steel ferrous oxide evolves in a free state on the boundaries between the grains and discontinuity appears in cast metal, especially when the temperature drops.

Table 2 illustrates the content of oxygen in steel during various melting stages and in steel ingots as determined by F. Edneral.

<div align="right">

Table 2
</div>

Oxygen Content in Steel in Per Cent by Weight

No. of heat	Before oxidised slag is skimmed	One hour after the beginning of reduction	Before pouring into ladle	In ladle	From ingot of first cheese	From ingot of second cheese
1	0.0070	0.0043	0.0034	—	0.0043	0.0049
2	0.0109	0.0022	0.0020	0.0024	0.0040	0.0046
3	0.0072	0.0030	0.0016	0.0044	0.0040	0.0048
4	0.0056	0.0032	0.0019	0.0030	0.0041	0.0048

When casting is made from complex steel alloyed with chromium and titanium it is necessary to pay attention to its oxidation and sometimes create artificially a neutral or reducing atmosphere of the surrounding gases during metal pouring. Oxygen decreases the solubility of carbon in austenite and ferrite and increases physical heterogeneity of pig iron and steel structure. The strength and hardness improve while ductility, yield point, impact strength and magnetic permeability decrease. Oxygen is believed to increase the susceptibility of steel to temper brittleness and corrosion and to defects during case-hardening. The presence of free ferrous oxides gives rise to hair cracks, banding and other flaws in rolled metal.

As the steel is deoxidised the amount of ferrous oxide decreases. Some of the oxygen passes into other oxides, MnO, SiO_2,

Al$_2$O$_3$, which are present in the steel in the form of dispersed nonmetallic inclusions.

The oxygen content in steel depends on carbon concentration

$$K' = \frac{[\% \; C] \; [\% \; O]}{P_{CO}}. \tag{6}$$

When at 1600° C the pressure of CO is 1 atm (abs) the coefficient K' changes between 0.0020 and 0.0025. When the CO pressure goes down the value of K' remains constant when the product [% C] [% O] decreases.

Steel intended for castings should be thoroughly deoxidised. The steel stops boiling when the concentration of ferrous oxide reaches 0.03 per cent, i. e., 6 to 60 times more than follows from the equilibrium constant of the reaction

$$FeO + C \rightleftharpoons Fe + CO! \tag{7}$$
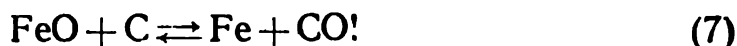
If the steel solidifies when the FeO content exceeds the equilibrium amount it begins to boil because of disturbed equilibrium and CO evolves again. In the course of cooling and solidification the solubility of FeO drops sharply, as well as the equilibrium constant, i. e., the reaction becomes more vigorous.

To prevent the evolution of carbonic oxide in the form of bubbles as the metal solidifies, the steel should contain at least five times more silicon than FeO.

Medium-carbon steels should contain at least 0.35 per cent silicon in order not to boil in the mould. Manganese alone is not enough to kill the steel although, according to equilibrium conditions, the theoretical amount of FeO residue with 0.50 per cent Mn comprises 0.2 per cent at a temperature of 1590° C.

Aluminium, the strongest deoxidiser in ordinary heats, can increase the content of free ferrous oxide in steel to 0.002 per cent and, given favourable conditions, even to 0.001 per cent (Fig. 11).

On cooling, liquid steel supersaturated with gases evolves bubbles which form gas cavities in the castings. If such blowholes develop at the surface their cavities are coloured dark while blowholes below the surface have smooth and light-coloured walls. This feature distinguishes them from shrinkage cavities with rough coarse-grained surfaces.

Gas cavities range from small blowholes hardly visible to the naked eye to large spheroidal or egg-shaped blisters. Pinholes occur mostly in small steel castings but sometimes can be found in large castings. The reasons for the formation of pinholes were extensively studied by Sims and Zappfe of the U. S. A., who confirmed Müller's surmise he put forward in 1882 to the effect that pinholes were caused by the precipitation of hydrogen from the advancing layer of freezing steel. Batty (U. S. A.) attributed his phenomenon to an additional absorption by the metal of

gases given off by the foundry mould during the early stage of freezing.

Blowholes appear when water vapours begin to evolve near the hard skin of a casting. The holes increase in size concurrently with the growth of the crystals (Fig. 12) because of the diffusion of hydrogen into the hole cavity from the surrounding liquid steel which begins to solidify.

When the molten steel contacts the moist walls of the mould it forms a hard skin and heats the working surface of the sand
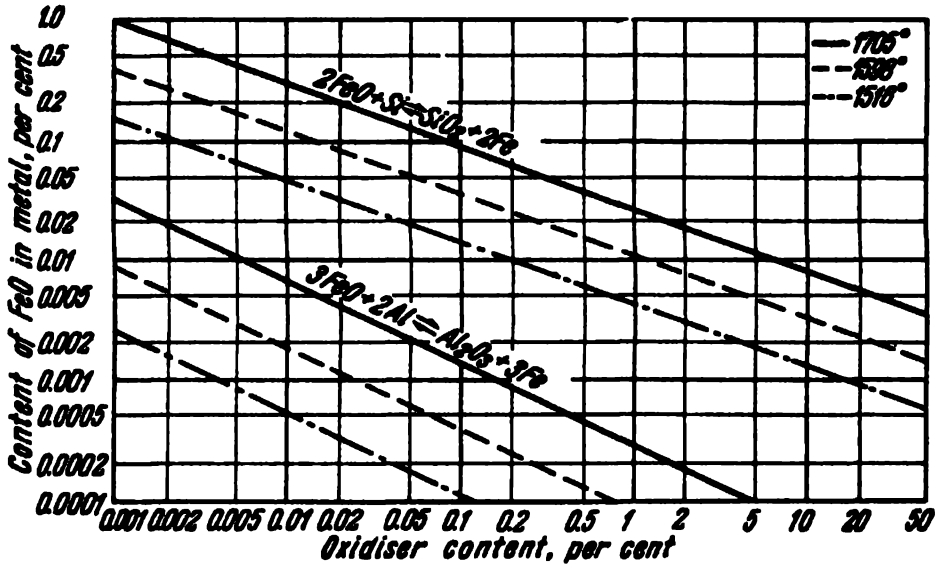
Fig. 11. Change in the amount of ferrous oxide in steel as a function of deoxidiser content in equilibrium conditions (Chipman)

mould. The water vapours in the layer of sand expand to about 5,000 of their volume, thereby building up an oxidising atmosphere.

Some vapours penetrate through the sand pores, contact the hot steel and form ferrous oxide and hydrogen by the following reaction:

$$Fe + H_2O \rightleftharpoons FeO + 2H. \tag{8}$$

Ferrous oxide forms an oxide film on the casting. Atomic hydrogen diffuses at a high rate at the near-solidus temperature and penetrates the oxide film to enter the solid steel while the remaining water vapours are ejected through the sand pores. The concentration of hydrogen in steel increases and, at a sufficiently high level, it evolves in a molecular form.

According to Sieverts, at 1 atm (abs) hydrogen dissolves in iron, as shown in Fig. 13, the solubility dropping especially sharply during solidification. When hydrogen dissolves in molten

iron no gas is evolved which can be attributed to its ability to produce supersaturated solutions when the solubility drops at a lower temperature. Like nitrogen, hydrogen cannot cause the steel to boil. But hydrogen may escape from a supersaturated solution due to a sharp decrease in the solubility during rapid transition from a liquid into a solid state. Knipp pointed out the reversibility of the reaction (8) which may produce hydrogen or water vapours. The rightward reaction is a weakly exothermal one; as the temperature drops the water vapours are reduced to hydrogen.
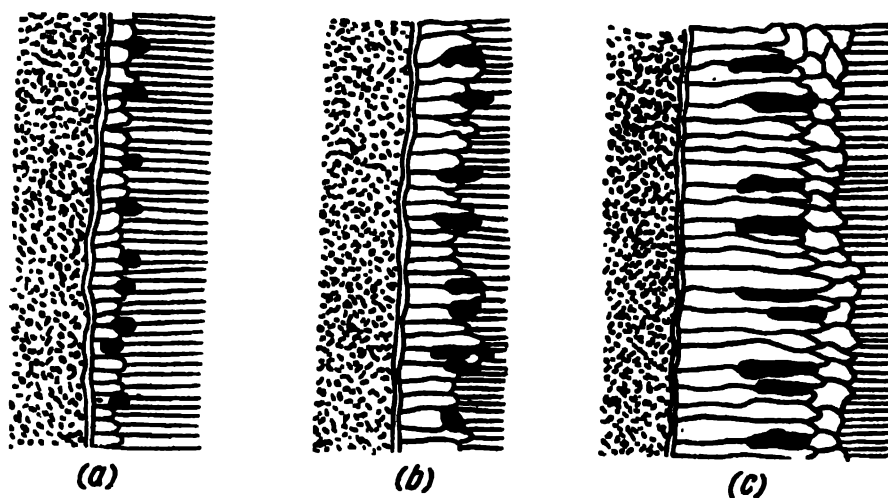
Fig. 12. Formation of subskin blowholes in castings (Sims, Zappfe):

a—initial stage; *b*—intermediate stage; *c*—final stage

The same effect is achieved by an addition of surplus aluminium which begins to react with the water vapours of the mould linings:

$$2Al + 3H_2O = Al_2O_3 + 6H \qquad (9)$$
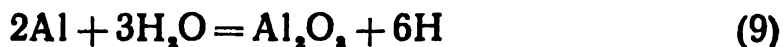
which we have observed in the experiments when aluminium was added to liquid pig iron containing much carbon and where there should be no ferrous oxide in a state of equilibrium. Incidentally, N. Dobrokhotov recommends an addition of 2 kg of aluminium to 1 ton of carbon steel.

When blowholes below the surface are opened they reveal thin slag films caused by an oxidising atmosphere which promotes the formation of oxides. The evolved gas bubbles get stuck in the tiny spaces between the ends of the dendrites. In some cases when, for example, the mould is excessively moist, individual blowholes become continuous gas cavities which pierce almost the entire casting.

Radtke experimented with casting thin steel sheets (Fig. 14). The left external halves were dry and the right halves—green-sand cast. A green sand square core was inserted into the right sheet.

The right halves of the left sheets and the square core in the right sheet were affected by pinholes. The boundaries between the sound and affected parts were sharply delineated. Despite the high solidification rate of these sheets only 10 mm thick, hydrogen had time to diffuse from the green-sand mould through the hard skin and holes developed below the surface.

To make sure that hydrogen was responsible for the formation of the holes, Radtke added 3 per cent sodium hydrate to the moulding sand mixture and obtained holes in the castings in places which directly adjoined the sections containing the mixture with the sodium hydrate. The same result was obtained when slaked
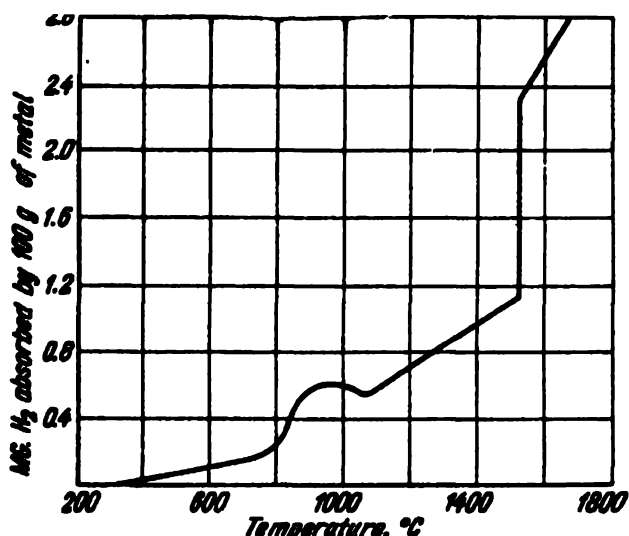
Fig. 13. Change in the solubility of hydrogen in iron as a function of temperature (Sieverts)

lime and other substances reducing the surface tension of the oxide films of liquid steel were added. Acidic additions and ferric hydrates combined with the lining mixture produced no pinholes in the castings because the surface tension remained high.

To prevent the formation of holes below the surface of steel castings, Czechoslovak metallurgists spray the surface of green-sand moulds with cement. The cement absorbs the moisture, added mechanically to the moulding sand mixture, and binds it chemically causing it to evaporate slowly when the surface of the mould contacts molten metal.

When the concentration of gas in steel is very high the pinholes grow in size and spread over the entire casting volume, which can be easily checked with the help of a test sample poured into a heated metal cup. The steel grows and forms a convex meniscus. An excessively moist surface of the mould or the core, moisture condensed on the surface of the chill or chaplet may cause large gas blisters to appear in steel castings. It happens sometimes that water vapours or air trapped in the mould break through into the cast metal which boils and even splashes out from the mould. One can easily explain these mechanical causes for the formation

of gas holes and prevent them by carefully observing the pertinent instructions.

According to A. Samarin and L. Novik, the normal gas saturation of basic electrical steel amounted to 13-60 cu cm per 100 g of metal after the hard charge had been melted, 23-65 cu cm at the end of the oxidation period and 5-14 cu cm before tapping. The hydrogen content in ready open-hearth or electrical steel usually comes to 0.004-0.005 per cent.

A decrease in the content of hydrogen from 7-9 to 4-5 cu cm per 100 g of liquid steel considerably increases the ductility of cast steel and prevents the formation of flakes and other defects. The
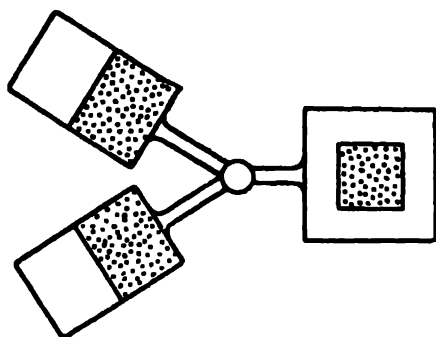
Fig. 14. Radtke experiment

charge should be free from rust formed of ferric hydrate $Fe_m(OH)_n$; only dry ore must be used and ferroalloys calcinated before being charged into the furnace, especially ferrosilicon; the ladles must be dried with the bottom up. Lime which has absorbed moisture and has been partially transformed from CaO into $Ca(OH)_2$ should never be added. It is good practice to charge chalkstone calcinated at 500-700° C to dry and remove constitutional moisture. All this was practised at the Kuznetsk Iron and Steel Plant in the U. S. S. R. where sound castings of improved quality have been obtained.

It should be noted in conclusion that hydrogen is one of the most harmful admixtures in steel used for the production of shaped castings.

Nitrogen apparently forms chemical compounds with iron:

$$\left.\begin{array}{l} 2Fe + {}^1/_2 N_2 = Fe_2N - 3{,}940 \text{ kcal} \\ 4Fe + {}^1/_2 N_2 = Fe_4N - 4{,}460 \text{ kcal} \end{array}\right\} \quad (10)$$
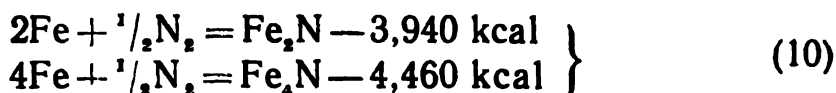
The Fe_2N nitride decomposes, as has been shown by N. Chizhevsky, in pure nitrogen at about 550° C. At higher temperatures, another nitride, Fe_4N, also undergoes dissociation.

The solubility of nitrogen in solid iron, according to the data collected by N. Morozov, is illustrated in Table 3.

The usual nitrogen content in steel, as determined by the method of hot extraction in vacuum, depends on the melting technique.

Table 3

Solubility of Nitrogen in Solid Iron

Phase	Temperature, °C	Solubility of nitrogen in Iron	
		mg/100g	per cent
α-iron	800	0.8	0.001
	900	1.6	0.002
γ-iron	900	21.6	0.027
	1400	16.8	0.021
δ-iron	1400	7.6	0.0095
	1530	10.4	0.013

Acid and basic electrical steel contains from 0.003 to 0.006 per cent nitrogen, open-hearth steel—from 0.002 to 0.004 per cent, Bessemer side-blown steel—from 0.003 to 0.006 per cent, and Thomas bottom-blown steel—from 0.007 to 0.020 per cent. After the steel has been treated with oxygen on the surface (the water cooling pipe is directed vertically through the mouth and almost reaches the bath surface) in a converter body with a basic lining, nitrogen content has been found at no more than 0.003 per cent. Such steel is by no means inferior to open-hearth steel by its mechanical properties. Nitrogen was believed responsible for the formation of stone-like fracture in steel and hydrogen—for coarse-grain macrostructure. Later, it was found, however, that the steel contained an increased amount of nitrogen because too much aluminium had been added.

A. Samarin, M. Korolyov and I. Paisov studied this problem and came to the conclusion that when nitrogen ferrochrome is added to stainless steel with 23 per cent Cr, the steel structure changes into close-grained, i. e., nitrogen proves to be a useful admixture in such steel.

The harmful effect of nitrogen in cast steel is eliminated by introducing some ferrotitanium into it before tapping in order to bind nitrogen into stable titanium nitrides.

9. CASTING DEFECTS RELATED TO MOULDING SANDS

The formation of gas cavities in steel castings depends mainly on the structure of the mould and the composition of the facing material. Vapours and gases contained between the grains of loam sand mixtures, and also in the mixture composed of sand and organic binders, condense on the surface of a green-sand mould, on chaplets, chills and washes and are the main causes for the formation of cavities. Surplus moisture on the surface is highly objec-

tionable. The air trapped in closed parts of the mould during pouring expands due to heating and, being unable to escape, pierces the weak surface film of the casting and may even destroy it completely by gas cavities.

When a loam-sand mould is filled rapidly with a superheated metal, the thin surface layer of the gating system and the mould cavity are quickly raised in temperature.

Let us consider the moment when molten metal (Fig. 15) rising from below has just inundated part of the mould wall and heated a layer h thick to 100° C. A large number of points
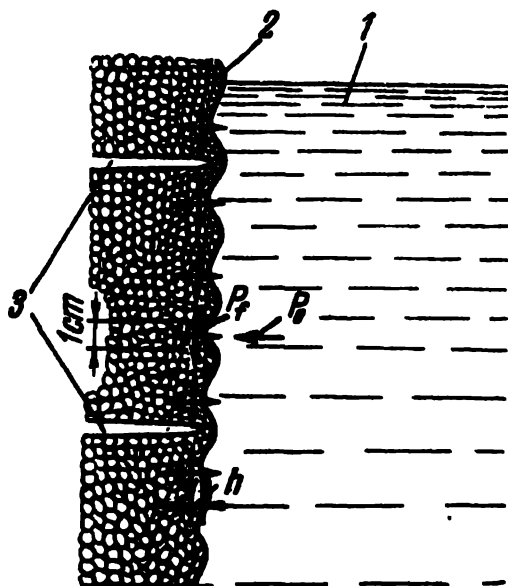
Fig. 15. Movement of vapour and gas in the subskin layer of a foundry mould (Bidulya):

1—liquid metal; 2—surface layer of the mould; 3—ventilation ducts

spring up, around which the mechanically introduced water begins to boil, thereby generating a high pressure vapour. These points are the pressure epicentres which develop on the isothermal surface at a certain distance from the sand-metal interface. The pressure of the vapour and gas begins to propagate like a ball wave.

Of any vectors of a ball, with pressure arising in its centre, it suffices to examine only four vectors in the plane of the drawing directed towards all the four sides. The vertical vectors are mutually neutralised by the adjacent opposing vectors. In other words, the movement of the vapour and gas will be essential either to the right from the isothermal surface towards the casting (let us call it rightward vector) or to the left, into the mould (leftward vector).

We shall first consider the rightward movement, and then the leftward movement. Each is possible in the presence of intergranular communicating pores forming capillaries in the facing. Each movement can be regarded as a filtration of gas through a porous medium which was the subject of close study by the hydraulics specialists of the first half of the 19th century—Slichter, Zunker,

Darcy and others. The flow of gas by weight through a porous filter can be found from the equation:

$$G = \frac{Fd^2(P_f^2 - P_0^2)}{192\mu\,KR\,\theta h},\qquad(11)$$

where G = quantity of gases by weight which filter through a porous medium in a unit time, g/sec;

F = area of the porous filter, sq cm;

d = mean diameter of the filter meshes, cm;

P_f = pressure of the filtering gases in the epicentres, g/cm^2;

P_0 = counterpressure which, in our case, may be taken as equal to the pressure head of liquid metal against the mould wall, g/cm^2;

μ = coefficient of hydraulic resistance for gases passing through the pores;

K = Darcy filtration factor accounting for the filter porosity according to Zunker;

$$K = a\,\frac{md^2}{\mu\,(1-m)^2};\qquad(12)$$

R = gas constant;

θ = absolute temperature, ° K;

$$m = \frac{V_1 - V_2}{V_1} = \text{filter meshes;}$$

V_1 = filter volume (grains + meshes), cu cm;

V_2 = volume of pores in the filter, cu cm;

h = filter thickness, cm.

Let us cut out a tube 1 sq cm in area in the filter and substitute the right side of Eq. (12) into Eq. (11) to obtain

$$G = \frac{P_f^2 - P_0^2}{192a\,\dfrac{m}{(1-m)^2}\,R\theta h}.\qquad(13)$$

This equation will help us to examine the filtration kinetics. The following three cases are possible:

1st case. $P_f > P_0$. The pressure of the gases in the mould exceeds the pressure head of the liquid metal against the mould wall (it is assumed that there is no hard skin yet on the casting). The gases break through the film on the casting surface, enter it and form gas cavities. If P_f considerably exceeds P_0, there may be an explosion and the liquid metal will sputter out from the mould.

2nd case. $P_f < P_0$. A dry-sand mould is filled in a vertical position. The hydrostatic pressure head of the liquid metal over-

comes the insignificant pressure of the gases in the mould and the liquid metal, as well as fusible oxides, rushes into the intergranular spaces of the mould surface forming what is known as metal penetration.

3rd case. $P_f = P_0$. The pressure of the gases in the mould and the pressure head exerted by the metal being cast on the mould are equalised. This is obviously the most desirable case, since it prevents the formation of gas cavities (1st case) and metal penetration (2nd case).

H. Dietert and his colleagues put forward a theory on the equalisation of gases on the surface of a foundry mould but he presented his theory in exactly the same way as Soviet scientists back in 1941, 1948 and 1955.

The analysis of Eq. (13) shows that to decrease the amount of the gases G filtering into the casting from the surface of the mould it is essential to reduce all the values in the numerator and increase them in the denominator. The difference in the numerator $P_f^2 - P_0^2$ can be represented as the product of $(P_f + P_0)(P_f - P_0)$. Quite obviously, P_f and P_0 should tend by their absolute values to minimum, or, in other words, the mould before pouring should be either dried or heated to a temperature above 100° C to bring the amount of water vapours to the smallest possible amount. To reduce P_0, the mould should be filled lying on its side or in an inclined position. If $P_f = P_0$, the difference converts to zero as does the quantity of the gases filtering towards the mould. In other words, *everything must be done to equalise the pressure P_f and P_0 on the metal-sand interface.* It also follows from this that as P_f is not equal to zero, P_0 should be small but still differ from zero. Thus, it is not altogether necessary that the moulds be completely dried or calcinated until they lose, not so much mechanical, but colloidal and solidification moisture.

The magnitudes in the denominator of Eq. (13) must be increased. The coefficient m expresses the ratio between the filter volume, minus the volume of the pores, and the filter volume together with the volume of the grains and pores and can be increased if the volume of pores is reduced to a minimum, i. e., the filter must be packed tight. According to L. Leibenzon, who developed the theory of medium porosity, loose packing of the filter round grains with pores in the form of spherical squares (Fig. 16) can be transformed into a tight packing with pores in the form of spherical triangles. However, the filter (thin surface layer of the mould) can also be packed, for example, by coating the mould surface with dry parting compound (silica or zirconium flour), with washes made of dispersed particles (silica and chrome-magnesite flour, chromite flour, ground graphite, peat dust). It is important that the compound or wash should be such as to nearly close the intergranular pores on the mould surface.

The expression $(1-m)^2$ enters the numerator of Eq. (13). Hence, to retard filtration, m must become a magnitude close to unity $(1-m)\rightarrow0$, i. e., the filter at the mould surface must be the tightest and least permeable for the gases. In other words, the permeability of this portion of the facing should be such as to prevent completely the passage of gases from the mould into the cast metal as well as the passage of superheated liquid metal through the mould pores.

The product of $R\theta=P_fV_2$ (Clapeyron equation) is the next term in the denominator of Eq. (13). P_fV_2 can be regarded as the work done by the filtering gas, and consists apparently of recurrent expansions and compressions of the elementary streams of gas
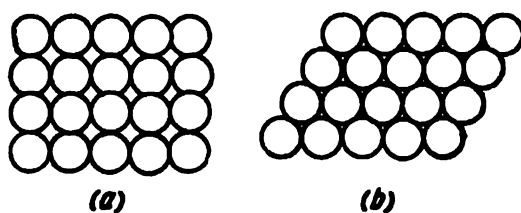
Fig. 16. Grains packed in a moulding
sand mixture:
a—loose; b—tight

which pass through the meandering pores of the filter with a variable cross section in the form of spherical triangles. The more the stream curves as it bypasses each ball-shaped grain of the filter (Fig. 16), i. e., the more grains there are per unit length of the gas path, the greater will be the work done by the gas and the filtration retardation. Hence, as before, the filter meshes should be as small as possible.

Let us now examine the last factor in the denominator of Eq. (13)—the value h, which is the filter thickness, very important for obtaining sound castings.

At first, the cold surface of the mould is heated by the incoming liquid metal. At this moment $h=0$, and all the gases that form either evolve mixing with the air, which fills the mould cavity, or escape through the overflows and the holes in the open risers. Provision should be made for the evacuation of air, gases and vapours especially from the closed internal cavities of the mould, as for instance, from the upper sections of the bent web of a flywheel through the overflows (Fig. 17) for, otherwise, the remaining air or gas will penetrate into the casting and spoil it. In foundry practice there is to this effect a current term "a good streamlined mould" suggested by P. Kosovsky. If the metal is poured in too fast it will immediately inundate the mould walls which have not been heated to a proper depth. When h equals nearly zero, the remaining moisture and gases will form a gaseous cushion at the

metal-sand interface if the mould surface is covered with such gas-generating substances as coal, coke or graphite. The gaseous cushion will be useful as long as the pressure P_f is counterbalanced by the pressure P_0. Otherwise, the gases will again break through the weak film on the metal and produce blowholes. There should be no hydrogen in this layer.

When the mould is filled with metal slowly, other conditions being equal, the filter will be thicker and the gases will encounter more resistance in piercing through the metal skin. And, finally, if the metal has formed a hard skin, as may be the case when the poured-in metal is superheated insignificantly above the temperature θ_s or when frozen moulds are filled, the gases will no longer
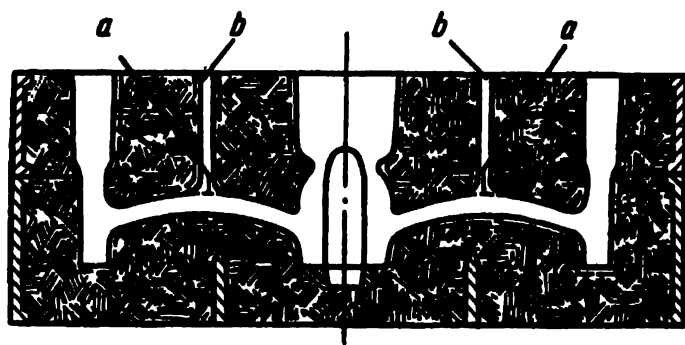
Fig. 17. Formation of air pockets *a* in a mould for a flywheel without overflows *b* (Bidulya)

be able to break into the metal being cast. Hence the conclusion that an excessive speed of filling is harmful rather than useful.

Let us consider now the movement of the gases from the pressure epicentres into the mould. Designated as the leftward movement, it is essentially the same as the rightward movement. But, in conformity with Eq. (13), the quantity of the evacuated gases should be preferably increased. In other words, the numerator must be increased or denominator decreased.

Since the difference between the squared pressures should in this case be greater, both the sum $P_f + P_0$ and the difference $P_f - P_0$ are also greater. Obviously, it is more advantageous to increase the pressure P_f and not P_0, i. e., speaking technically, the mould or the core should not be completely dried or the pressure P_f brought down to zero. Let us note in passing that, as the gases and vapours move to the left from the isothermal surface, they pass through the cold capillaries of the filter. If the amount of the filtered water vapours will be excessive, the condensate which settles on the cold walls of the mould will close the channels and instead of proceeding leftwards the gases will be vented to the right overcoming P_0, i. e., they will break into the cast metal. And, conversely, if the gas permeability of the filter to the left from the

isothermal surface is sifficient, the condensate will not clog the channels and filtering will proceed in the leftward direction.

We shall arrive at the same conclusion if the filter density m in the denominator is greater, i. e., if it differs from unity (absolutely dense filter). Quite the reverse result will be obtained when this layer of the mould is tightly packed as, for example, in a very tight packing of both the thin surface layer and the entire facing by violent blows with a rammer and prolonged jolting in a moulding machine. It is a well-known fact that a vigorous compacting of a mould noticeably increases the number of blowholes in the castings and makes them completely unfit for use.

The escape of the gases $P_f V_2$ to the left should be facilitated by using coarse-grain mixtures and providing for gas outlet holes, gas-permeable voids in the central part of the mould or core and holes in the moulding box walls to remove gases from the mould.

The kinetics of the gas and vapour filtration through the mould pores suggests the following measures for the prevention of casting defects related to moulding sands.

1. The thin surface layer of a sand mould or core should have a dense and refined structure which can be achieved by spraying the surfaces of a green-sand mould with dry powders composed of dispersed particles generating no gas. Here belong silica flour, peat dust, ground chromite, portland cement, ground chamotte powder, zirconium flour, etc. Dry-sand moulds and cores are coated with a thin layer of wash made of these powders dissolved in sulphite tar or pure water. Dried mould or large cores are sometimes washed with water or swabbed again (when the mould has not yet cooled after drying) or rubbed. In all cases the cracks and pores developed during drying should be closed up to create a near-zero gas permeability on the surface of the mould or core.

2. Besides the gases entrapped in the metal, water vapours and volatile organic substances evolved in heating a green-sand mould or core are also responsible for the formation of gas cavities. Sound castings can be obtained when the pressures of the gases and vapours which escape from the mould walls during pouring are equalised. In each case measures should be taken to balance the gas pressure in the mould by decreasing this pressure, for example, by reducing the moisture content of the green-sand mixture, drying the moulds and cores, mainly on the surface, binding the moisture into colloidal compounds, as, for instance, in mixtures with bentonite or soluble glass, and properly positioning the mould during pouring, etc. Neither the mould nor core should be excessively wet.

3. It also follows from the above that there should be no condensed moisture on the working surfaces of the moulds and cores, on chills and chaplets.

4. The vapours and gases are evacuated from the subskin layer of the mould mainly through its pores. Every possible means should be employed for the purpose: filling sand made from coarse-grain materials such as open-grain sand and ground chamotte, ventilation channels and holes in the side walls of the moulding box.

A special attention must be paid to the ventilation of the mould bottom in floor moulding not only to prevent the formation of gas cavities but also to safeguard the servicing personnel.

5. When the pressures P_f and P_0 are equalised on the surface of the mould, it no longer needs to be dried. Prior to filling, large moulds should be dried on the surface to a small depth specified in each case.
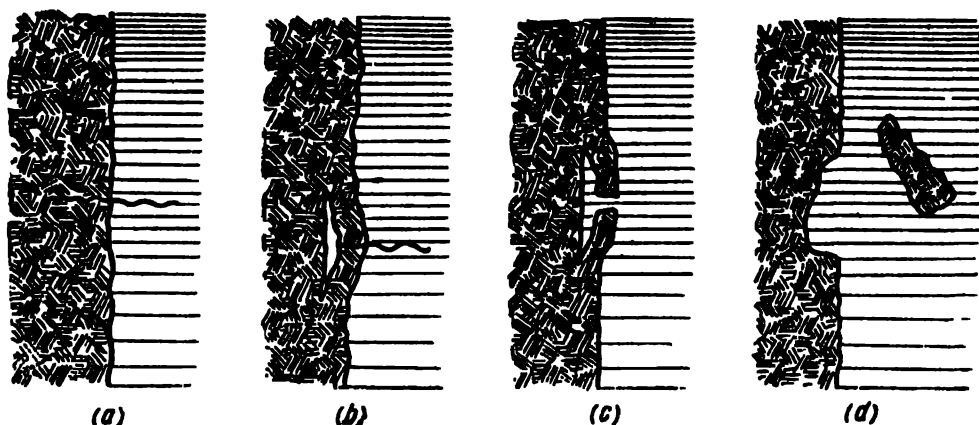
Fig. 18. Defects caused by the moisture boiling in the subskin (Bidulya):

a—cracks in the mould and casting; *b*—bulge in the mould, and crush and crack in the casting; *c*—bulge and crack in the mould and veining in the casting; *d*—drop in the mould

6. Filling of the moulds with a highly superheated metal brings about the same consequences as the excessively high pouring rate. As follows from the further exposition, the interval of pouring temperatures should be specified for each case, being dependent not only on the grade of steel but also on the size of the casting.

7. The gases and vapours are evacuated from the closed parts of the mould during pouring through channels (overflows) arranged at the top of every blind mould cavity.

8. Core prints should be tightly stowed in their recesses to prevent the liquid metal which broke through the gaps from clogging the gas outlet channels issuing from the core and the mould.

The subskin pressure of gases in the mould or core, which develops during pouring, gives rise not only to blowholes in the casting but to other defects as well, for example, veining. A lot

of time and effort was given to elucidate the causes for veining formation.

When the mould surface is heated by the liquid metal and when the moisture content is too high the water may begin to boil in the subskin layer producing cracks and bulging in the mould facing (Fig. 18).

A drop in a mould is a defect caused by the breaking away of part of the moulding sand, which mixes with the metal, by the heat of liquid steel at too slow a pouring. The defect can be prevented by increasing the pouring rates, drying the moulds for heavy casting to a depth of at least 200-300 mm and sprigging the mould face with nails and iron hooks.

10. THE EFFECT OF NONMETALLIC INCLUSIONS ON CAST STEEL

Steel and other alloys are never quite pure. They always contain gaseous and other impurities known as nonmetallic inclusions, such as oxides, sulphides, carbides, phosphides, nitrides and many other still more complex nonmetallics associated with metals.

Few of them can be regarded as useful admixtures. For exapmle, under certain conditions, carbides may increase the strength of steel at high temperatures, but most nonmetallic inclusions are rather harmful because they sharply reduce the ductility and strength of steel.

The degree to which nonmetallic inclusions affect the steel depends on their amount, shape and distribution in macro- and microstructures.

More dispersed inclusions having a greater contact area with the metal grains decrease the ductility and strength much more noticeably because of a weaker bond between the metal and the inclusions. The greatest harm accrues obviously from dispersed inclusions yielded by the reactions which occur in the molten metal. They cannot float out of the steel because of a short-life nature of these reactions and get stuck in the liquid metal.

Nonmetallic inclusions 'entrapped in the steel mechanically as, for example, slag particles mixed with metal during ladling, the material of a washed-out lining, spout, ladle or the channels in the foundry mould, particles of an eroded sand mould and as well as other alien admixtures which enter the steel from outside, are larger than the type discussed above. They deteriorate mainly the strength rather than the ductility of steel. When the steel is soaked for some time prior to becoming solid these inclusions may float out to the surface of the metal cast and get stuck there during the process of solidification. They can be eliminated by removing the upper portions of castings in the form of risers or specially provided extra stock.

The movement of admixtures in liquid metals is governed by Stokes's law. The rate of motion is determined as the difference between the density of the metal and the admixture. During contraction, a shrinkage cavity and a vacuum develop in the massive parts of the ingot or casting, especially in the risers. The inclusions are now affected in their motion not only by the gravity force but also by the pressure difference. The inclusions rise under the pressure difference along a curved path producing curved thread-like holes in the casting where the sulphur inclusions are trapped towards the end of solidification and form "whiskers".

The thread-like arrangement of inclusions has an adverse effect on the mechanical properties of cast steel. Closely grouped inclusions form clusters like the stars in the sky and are, therefore, called "galaxies". This is a eutectic type of arrangement. It also influences the properties of the metal and may cause the casting to break prematurely. Meshy inclusions which envelop and separate the grains of metal are most unwelcome, however. The strength of such structure depends on the strength of the material from which the inclusions are made and not of the metal.

Egg-shaped inclusions (as, for example, dendritic inclusions of sulphurous titanium in steel) as well as needle-like and sharp-edged impurities build up concentrated stresses and produce a negative effect on ductility.

The least harmful are spherical inclusions with a minimum surface at an equal volume. The balls may form in a molten metal when their melting point is slightly above the solidus temperature. The material of the inclusions should possess the maximum surface tension and the ability to coagulate, i. e., to agglutinate liquid dispersed particles.

The amount, shape and distribution of inclusions in steel is an important characteristic which predetermines the operating properties of cast metal. Bringing them under control opens up a new page in quality metallurgy and is no less important than the science of alloying, as it enables physically homogeneous metals to be cast and increases their mechanical properties almost infinitely, far in excess of the usual limits of static strength and ductility of steel both at ordinary and elevated temperatures.

Nonmetallic inclusions in steel owe their origin to the process of oxidation of iron and its admixtures as they are melted in the atmosphere of furnace gases containing oxygen.

Prior to tapping the steel is deoxidised by transferring the oxygen of the ferrous oxide to oxides of manganese, silicon, aluminium, etc. A thoroughly deoxidised steel contains residuals of iron oxides dissolved in the metal, some of which enter duplex, triplex and more complex oxides of the type $mFeO_1 \cdot nMe_1O \cdot pMe_2O$, where Me_1 and Me_2 are deoxidising agents.

The more complex is the oxide, the lower is its melting point. It is present therefore in liquid superheated steel in the form of spherical droplets, i. e., in the most desirable form.

For this reason, the composition of oxides is controlled by introducing several deoxidisers simultaneously, rather than one, into liquid steel. This is known as complex deoxidation.

The proper choice of a complex deoxidiser, for example, silico-manganese, silicocalcium, AMS alloy (aluminium, manganese, silicon) produces a metal of highest purity whose nonmetallics agglutinated in large drops pass partially into slag.

The mixing of steel and slag, as suggested by Tochinsky and Perrin makes the steel much more pure in the same way as complex deoxidation.

Pure oxides are rather rare in ordinary steels. For example, the silicates are coloured dark and the sulphides are light grey. The oxides and sulphides tend to mix together to form oxysulphides whose colour depends on the content of oxides or sulphides.

If the steel is deoxidised with aluminium, it develops oxysulphides of aluminium and iron, aluminium and manganese and complex silicates and aluminates. When used as a deoxider, titanium combines with iron to produce the oxide $m\mathrm{FeO} \cdot n\mathrm{TiO_2}$ or the mineral ilmenite. In chromium steels the oxysulphides contain CrO and $\mathrm{Cr_2O_3}$.

The silicates of manganese and iron in spherical or ellipsoidal form can be encountered in steels melted in furnaces with an acid lining. These steels are remarkable for their high impact strength especially when they contain little sulphur and phosphorus.

Today, the method of deoxidising steel with aluminium is being revised with due regard for a number of rules elaborated in the course of many years of practical experience.

When a small amount of aluminium is introduced into liquid steel nonmetallic inclusions precipitate in spherical form. Additions of aluminium of 0.025 to 0.5 per cent reduce the inclusions to the eutectic form and a "galaxy", depending on the sulphur content: if there is 0.03-0.05 per cent sulphur, very few inclusions of the eutectic type precipitate. With 0.075 per cent Al the inclusions are arranged in galactic clusters. An addition of 0.10 per cent Al distributes alumina inside the metal grains and the sulphides—on their boundaries. At 0.50 per cent Al the alumina is observed on the boundaries, enveloping the grains, which embrittles the metal. The aluminium sulphides are dark in colour.

Complex deoxidisers containing calcium with 0.10-0.20 per cent aluminium or titanium, or zirconium yield spherical duplex and triplex oxysulphides. To all appearance, these inclusions are duplex sulphides. Given the favourable shape of nonmetallics, the steel will possess sufficiently good mechanical

properties. The effect of aluminium additions on these properties has been ascertained during experiments the results of which are illustrated in Table 4.

Mechanical Properties of a Normalised   Medium-carbon Cast Steel * With and Without Aluminium Additions
(Sims and Lillieqvist)

	σ_b, kg/mm²	σ_s, kg/mm²	δ_{10}, %	ψ, %	Izod Impact, ft-lb
Dry-sand mould. No aluminium	57.0	34.0	31.2	54.6	37.0
Dry-sand mould. 0.05 per cent Al	57.2	35.8	26.7	40.1	31.0
Green-sand mould. 0.05 per cent Al	57.4	36.0	24.7	35.1	26.5

If a steel contains much iron oxide, the silicates and sulphides begin to evolve at an early stage of solidification in a spherical form and the ductility of steel is not deteriorated. Conversely, when there is little iron oxide the solubility of the sulphides increases and they appear later in a eutectic form thereby reducing the steel ductility. Hence the conclusion that the boiling of steel in a furnace should be such as to limit the carbon content to small values to increase the FeO content. Then the steel should be rapidly recarburised, and deoxidised prior to tapping With this aim in view, the U. S. foundrymen sometimes charge iron ore into the ladle to increase the ductility, but the green-sand castings are found to be affected by pinhole porosity. If a steel is deoxidised with an equal (by weight) amount of aluminium the ductility of the metal increases. The joint effect of sulphur and aluminium is illustrated in Fig. 19. The ductility decreases with a higher sulphur content. Aluminium produces almost no effect on the ductility of steel with 0.02 per cent sulphur. This can be accounted for by the fact that in the absence of aluminium additions, when small amounts of it are added, the inclusions take a spherical form and the steel has a high ductility. With eutectic inclusions the ductility is low. Large-size irregularly-shaped inclusions again increase the ductility. With an addition of 0.04 per cent Al the ductility is quite satisfactory. The best ductility is attained when the steel is deoxidised with 0.075 per cent aluminium. Any further increase in aluminium content will do more harm than good.

* 0.28% C, 0.37% Si, 0.79% Mn, 0.03% S and 0.035% P.

Additions of 0.05 per cent Zr and 0.05 per cent Ti to medium-carbon steel produce eutectic-type inclusions with the resulting decrease in the steel ductility.

Some 0.06 per cent Al and 0.05 per cent Ca added to nickel-manganese steel produce a good ductility as well as silicocalcium without aluminium. As deoxidisers, beryllium and zirconium can be effectively used instead of calcium. About 0.02 per cent Be is tantamount to 0.1 per cent Ca, while 0.01 per cent Zr to 0.01 per cent Ca.

Ordinary steel intended for castings should be preferably deoxidised before tapping by adding 0.1-0.2 per cent of an alloy of calcium with manganese and silicon and 0.05-0.1 per cent Al.
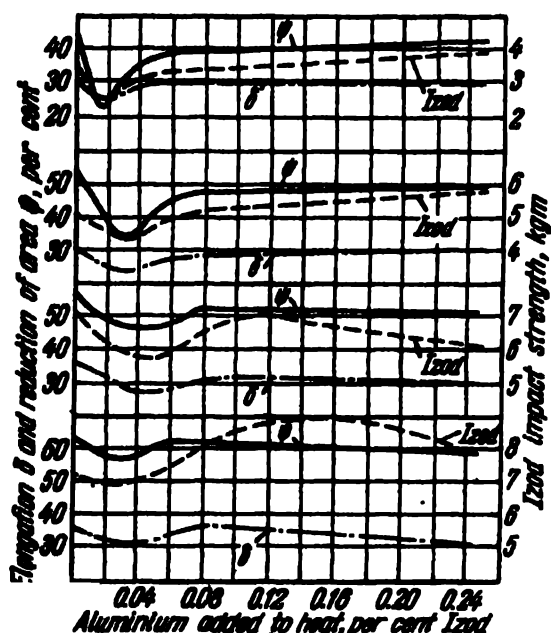
Fig. 19. The effect of aluminium on the properties of medium-carbon steels of various sulphur content (Sims and Dahle):
two upper areas on the diagram—over 0.05%; two bottom areas—about 0.02%

In the latter case, sulphides segregate in the form of large angular inclusions and tend to increase the steel ductility. The effect produced by a joint addition of calcium and aluminium is tantamount to 0.1 per cent Al. Calcium sulphides do not dissolve in liquid metal. But calcium alone deoxidises steel rather weakly and produces a much more better effect when used in combination with aluminium.

According to our data, an addition of 0.01 per cent Ca and 0.075-0.1 per cent Al yielded absolutely no precipitates during the ductility and impact strength tests.

No aluminium is used for deoxidation in the basic open-hearth process to prevent the formation of gas holes in dry-sand castings. In green-sand moulding the steel is deoxidised by introducing a small amount of aluminium (0.025-0.05 per cent) which in no way affects the ductility of steel as very little sulphur is left in the metal.

By blowing powdered deoxidisers at the end of heat into a basic open-hearth or electric arc furnace the metal will be almost completely divested of sulphur and additions of large amounts of aluminium will be no longer necessary.

Special conditions arise in producing heavy castings because test pieces have to be cut from full-sized sections during the metal tests. Low-temperature tempering is resorted to in such cases to prevent faults in the ductility of cast metal.

Under these conditions aluminium reduces the ductile properties of steel, just as titanium and vanadium. For this reason, such deoxidisers cannot be recommended for use.

When the thickness of the walls of cast test pieces is increased the ductility of steel drops sharply. The steel ductility is quite sufficient in heavy sections even without additions of these deoxidisers.

The experiments conducted in the U. S. A. on low-temperature ageing yielded the following mechanical properties of steel (Table 5).

Table 5

Ductility of Cast Steels in Heavy Sections of Castings (thickness 100-250 mm)

Amount of deoxidiser * per ton of steel	Ageing**	σ_b, kg/mm²	δ, %	ψ, %
None	No	} 71.5***	21.7	51.9
	Yes		22.3	51.9
2 kg of ferro-titanium	No	} 68.3	23.0	34.9
	Yes		25.7	48.4
0.8 kg of aluminium	No	} 68.8	20.2	25.4
	Yes		23.3	41.4
0.10 per cent V	No	} 76.0	17.5	39.6
	Yes		19.0	56.1
0.8 kg of CaMnSi	No	} 60.8	26.8	57.3
	Yes		27.2	62.6
0.8 kg of CaMnSi and 1 kg of aluminium	No	} 61.5	22.8	39.8
	Yes		26.3	50.3

* Ladle deoxidation.
** Heating at 200°C for 16 hours.
*** A normalising heat treatment given.

Nonmetallic inclusions which find their way into the metal mechanically are much larger in size than the nonmetallic products of the deoxidation reactions. Such inclusions are very danger-

ous *because they may cause sudden failure of cast machine* elements *in operation.*

The greatest danger, however, accrues from the slag of basic steel which remains in a liquid state at the pouring temperature (1550-1650° C). Acid slag begins to solidify at such temperatures. But when the steel is poured through the sleeve in the bottom of the ladle the slag does not enter the mould since it floats on the surface of the molten metal. Acid steel from small electric furnaces (up to 10 tons) is poured through the ladle spout which prevents the solidified slag from entering the mould with the metal. Hence, no slag traps are required in the gating system for steel castings. It would be a shear waste of liquid metal. The metal for steel castings (see p. 141) should be fed from the pouring basin to the mould cavity over the shortest path possible.

The degree to which particles of sand or any other refractory material detach from the gating channels is proportional to the kinetic force of the moving metal. In other words, the work expended on the erosion of the channel walls is proportional to the kinetic force:

$$A = \frac{mv^2}{2},\tag{14}$$

where A = kinetic force of the moving metal, g-cm/sec;

$\quad m$ = moving metal, g;

$\quad v$ = linear velocity of movement, cm/sec.

Replacing v by $\mu\sqrt{2gH}$,

we obtain after cancellation

$$A = \mu^2 mgH,\tag{15}$$

where μ = hydraulic resistance;

$\quad g$ = gravity acceleration, cm/sec²;

$\quad H$ = pressure head, cm.

This equation shows that to protect the channel walls against erosion the value of A should be reduced to a minimum, hence, also all the terms in the right-hand side.

In order not to contaminate the metal being cast by mechanically disintegrated particles from the gating system walls it is necessary:

(1) to reduce the amount of the moving metal and deliver it to the mould cavity through several, rather than one, ingates;

(2) to keep the mould during pouring in a horizontal or inclined position to reduce the pressure.

For steel casting the squared value of μ in Eq. (15) is properly reduced not by the "retarding elements" suggested by V. Fundator in the thirties but by bottom gates.

This system has noticeable advantages as a "self-retarding" type, as can be easily seen in Fig. 20. The liquid flows along the bottom channels as through a "flooded" opening. When the liquid in the left vessel (pouring cup at the beginning of pouring) is at its highest level it moves with a pressure head equal to the difference in the levels in both the communicating vessels (the right-hand vessel is the mould cavity empty at the beginning of pouring) at the full pressure head. During pouring the level of metal in the left-hand vessel is at a constant height while it rises gradually in the right-hand vessel. The difference in the levels is $h_1 = h_2 \to 0$. Hence, the probability of eroding the channel walls
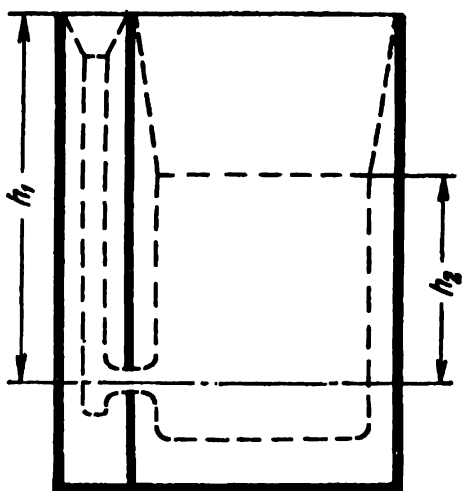
Fig. 20. Diagram of hydraulic retardation of a bottom gate (Bidulya)
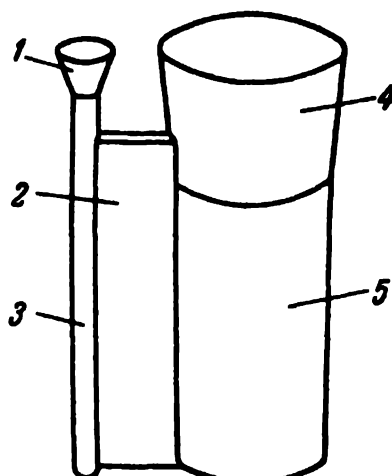
Fig. 21. Vertical knife gate for a steel casting (Bidulya):
1—pouring cup; 2—ingate; 3—sprue; 4—riser; 5—casting

will be the greatest only at the beginning of filling; then the linear velocity will gradually decrease to zero when $h_1 - h_2 = 0$. It can be assumed therefore that the velocity and the work of eroding the gating channels are gradually brought down to zero.

The advantages of a bottom gate over top gate and shower gate systems, etc., are quite obvious.

The same features are inherent in vertical knife gates (Fig. 21) which proved very effective in the ferrous metal foundry practice. They are distinguished by the same pressure head, linear velocity of movement and almost identical total hydraulic resistance as bottom gates.

To obtain heavy castings and reduce at the same time the erosion of the gating channels passing steel at a very high rate, the moving metal should be obviously separated into as many small streams as possible, i. e., provision should be made for as many ingates as there are thick sections in the casting.

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Although solidification of pure metals or binary alloys has been the subject of research by many investigators, we still lack a generally recognised theory of solidification. Only separate aspects of this theory have been elaborated with various degrees of trustworthiness. This fact can be attributed not only to the extremely complex phenomena occurring during solidification, which depends on several constant values, but also to the state of a liquid alloy prior to solidifying with its chemical and physical heterogeneity.

As distinct from the Van der Waals theory, prevalent for a long time, on the similarity between a liquid and gaseous states, the Soviet physicist Y. Frenkel showed the molecular structure of liquid and explained the features common to a liquid and solid state.

When a thin beam of X-rays is passed through a crystal, the film reveals a system of diffraction rings. If, under the same conditions, the rays are passed through a liquid, one or several diffused rings can be observed. With the temperature of the liquid approaching the melting point, the rings grow in sharpness and their number and position change. This indicates to the similarity in the structure of a liquid and a solid body. Further investigations of the intensity with which X-rays are scattered by liquids have shown that the latter attain a certain degree of orderliness at the temperatures near the melting point. It has been found that melting disturbs only the so-called long-range order. The short-range order in the atomic arrangement is preserved, i. e., the liquid is characterised by a certain orderliness in the atomic arrangement very close to the order in the arrangement of atoms in a solid crystalline body. The same result will be obtained if we compare the heats of melting and evaporation, as well as heat capacities. For example, the heat of iron melting is only 65 kcal/kg, whereas its heat of evaporation runs into 1,600 kcal/kg. Hence the conclusion: when a molten metal passes into a gaseous state the interatomic bindings are completely eliminated, but when a solid body is melted they are only slightly weakened. The heat capacity of iron in a liquid and solid state equals 0.17 and 0.19 kcal/kg °C, respectively, the values very close to each other. This also confirms our thesis.

As the temperature decreases, the magnitude of the oscillating motion of atoms goes down. A drop in the oscillating amplitude to 12 per cent of the interatomic distance will induce a solidification process. A solid metal consists of crystals which are

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* This Section is written by V. Saveiko, Candidate of Technical Sciences.

characterised by a definite regularity in the atomic arrangement in conformity with the space lattice typical of the given metal.

However, a decrease in temperature alone is clearly not enough to start the process of solidification. The nucleus of a crystal is needed.

In 1928, G. Stewart introduced the concept of the so-called sibotaxis blocks, meaning very small groups of molecules with a crystal structure which are in a dynamic equilibrium with that part of the liquid which is in a disordered "really liquid" state.

According to P. Dankov, the solidification nucleus may be the phase with a crystal structure similar to that of the parent material with the difference in the parameters of the space lattices not exceeding 15 per cent.

On the basis of the recent research data N. Chvorinov distinguishes three types of crystal nuclei according to the mode of their appearance.

1. Nuclei which arise spontaneously upon strong supercooling of steel.

2. Nuclei which are forced to arise by active alien particles as, for example, oxide of aluminium, titanium, etc. Such nuclei may appear even at an insignificant supercooling. Their number depends both on the degree of supercooling and the amount of active particles.

3. Nuclei already present in the melt which were formed either from isomorphous crystallites or from incompletely dissolved other metals (additions), or from the unmolten remains of the parent metal. Their number does not depend on the degree of supercooling.

D. Chernov was the first to establish that liquid steel crystallises in the form of dendritic grains so that each crystal resembles an elongated octahedron. Chernov's idea on the growth of dendrites which form columnar and polyhedral grains served later for other investigators as the basis for their hypotheses on alloy solidification.

G. Tammann has shown that the polyhedral grain structure of metal is brought about by the growth of crystallites in the form of polyhedrons or spheres from the solidification nuclei. Experimenting with organic compounds he has proved the presence of certain dependences between the number of the nuclei, the rate of solidification and supercooling of liquid, known as Tammann's laws of solidification. His work gave rise to the theory of volumetric solidification, according to which all liquid metal poured into the mould is subjected to a strong supercooling and isolated crystals grow simultaneously throughout the entire volume, some of them precipitating to the bottom of the casting. The heat of solidification liberated during the growth of the-

crystals is expended to raise the supercooled metal to the solid-ification temperature.

The recent experiments conducted on radioisotopic tracers completely nullify the theory on the "rainfall" of crystals. At the same time, the investigations conducted by N. Gudtsov in the early thirties showed that no supercooling was observed in real conditions in solidifying ingots made of technical metals. This served the basis for the development of the theory of progressive solidification which is gaining today ever wider popularity.

According to this theory, isolated crystals do not grow in nonsupercooled metal. They can grow only at the mould surface or near the already solidified layers of metal to which they impart the heat of solidification.

The theory of progressive solidification accounts for the three-zonal structure of a steel casting (Fig. 22) as follows.

The roughness of a foundry mould causes local increases in the rate of cooling and spontaneous appearance of a large number of solidification nuclei. The crystals which grow from these nuclei have no time to develop and form the outer fine crystalline zone. These fine crystals form a net succeeded by an aggregate of columnar crystals (dendrites) of the second zone. The dendrites appear when the heat transfer to outside proceeds faster than the equalisation and decrease in the temperature of the liquid metal in the middle zone of the casting. If the heat transfer and the temperature drop inside the casting are equalised, the solidification process may be interrupted and part of the crystals dissolved (Gudtsov's theory of wavelike solidification). As the temperature decreases further stepwise, the crystallites will continue to grow. At a still lower temperature the crystals will arise throughout the entire central zone simultaneously, due to a uniform temperature distribution in this zone. Such crystals grow in all directions and their nucleation is random. This is the zone of equiaxed dendrites.

N. Chvorinov believes that the division of the structure into the zones of fine crystals and columnar dendrites is rather arbitrary. He asserts that the second zone is the natural continuation of the first zone with the crystals which, being situated for growth in a direction perpendicular to the mould wall, develop some
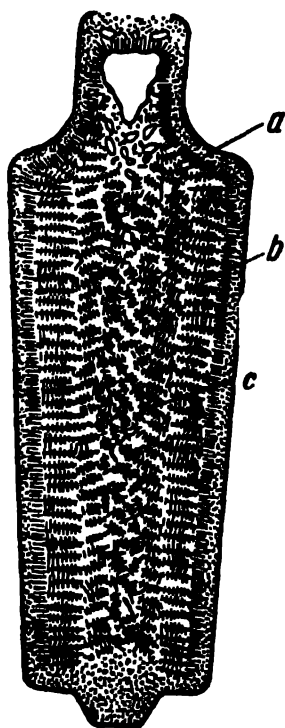

Fig.22. Section through a steel casting (Trubin, Oiks):

a—zone of globular crystals; b—zone of columnar crystals; c—zone of equiaxed crystals

distance in advance of other, not perpendicularly oriented crys-
tals and hold their growth in check.

A. Pronov has shown in his experiments that a crystal struc-
ture may arise as a result of the birth of new nuclei when liquid
metal flows in conditions of low superheating and, especially,
when the temperature of steel in the stream is reduced to the liq-
uidus temperature as, for example, when little superheated steel
is gated from the bottom. With a lower superheat and more in-
tensive movement of the metal in the mould, the solidification
nuclei will increase in number and the final structure will possess
a more refined pattern.

A fine crystalline structure may also form when liquid steel
is shaken in the chill mould under the effect of ultrasound oscilla-
tions or when a solidifying metal is stirred with a rabble. These
methods are not very widespread, however.

The higher is the rate of cooling of an alloy being cast, the
smaller the primary crystallites and shrinkage voids between them,
as well as gas bubbles, and the better are the mechanical proper-
ties of the casting.

Alongside the freezing rate, very important is the rate of solid-
ification which depends on time. In turn, the time during which
the casting solidifies (from the surface to the centre) depends on
many factors (and, apparently, first of all on the shape of the
casting). The solidification time is one of the most important
criteria which give true mechanical and operating properties of
castings. Many publications have been devoted to this question
in recent years.

One of the first was that of N. Chvorinov (Czechoslovakia)
who determined the time of full solidification of castings with
the aid of the Stefan square root law developed in 1890:

$$x = k\sqrt{\tau}, \tag{16}$$

where x = thickness of the solidified layer of metal, cm;
   τ = time of solidification, sec.

This law holds, however, for such casting practice when the
value of k remains constant. This can be observed in castings in
the form of an infinite flat plate solidifying in a sand mould,
i. e., at the minimum Biot number ($Bi = 0.1$). In other cases,
Eq. (16) will yield wrong results.

The solidification constant is usually determined experimen-
tally. But it can also be determined by calculations, true, rather
involved, which are based on the theory of similarity and heat
exchange elaborated by M. Kirpichov and A. Gukhman.

The theory of solidification was developed by N. Chvorinov
(1940) in Czechoslovakia and A. Veinik (1953) in the U.S.S.R.
N. Girshovich and Y. Nekhendzi have published a work where all
the solutions are of an approximate nature. However, in this

case too, calculations may prove helpful in determining the effect of the principal factors which describe the physical meaning of the phenomena occurring in the casting and the mould after it has been filled with molten metal.

To simplify the solution, A. Veinik suggested that the entire freezing process be broken into the following four stages:

(1) freezing of metal as the mould is being filled;
(2) transfer of the superheat from immobile metal;
(3) transfer of the heat of solidification;
(4) freezing of a completely solidified casting.

Since there is no casting in the first stage, its design temperature can be found only in the second stage proceeding from the equation of the heat balance of the casting: the amount of heat extracted from the surface of the casting is equal to the decrease in its heat content.

The equation for determining the heat transfer from the surface of the casting has the form:

$$dQ = \alpha_1 \left(\theta_1 - \theta_{2in}\right) F_1 \, d\tau, \qquad (17)$$

where dQ = amount of heat transferred from the surface of the castings, kcal;

$\alpha_1 = \dfrac{q_1}{\theta_1}$ is the coefficient of heat transfer from the surface of the casting;

q_1 = heat content of liquid metal, kcal/cu m;
θ_1 = mean temperature of liquid metal in the given cross section, °C;
Q_{2in} = initial temperature of the mould, °C;
F_1 = area of the casting, sq m;
$d\tau$ = given time interval, hr.

In this equation the arbitrary coefficient of heat transfer α is the specific heat flux as referred to the temperature drive and has nothing in common with the heat loss coefficient used in the heat exchange theory. In a general case

$$\alpha_1 = M \left(\tau - \tau_0\right)^{n-1}. \qquad (18)$$

The heat content of the casting will decrease during the time $d\tau$ by

$$dQ = -V_1 \gamma_1 c_1' \, d\theta, \qquad (19)$$

where $V_1 \gamma_1 c_1'$ = volume, specific weight and heat capacity of the casting respectively;

$d\theta$ = difference between the temperature of the liquid metal and the initial temperature of the mould, i. e., $\theta = \theta_1 - \theta_{2in}$.

Equating the right-hand sides of Eqs (17) and (19) we shall obtain the equation of the heat balance of the casting:

$$M \left(\tau - \tau_0\right)^{n-1} \theta F_1 \, d\tau = -V_1 \gamma_1 c_1' \, d\theta. \qquad (20)$$

After separating the variables we shall get

$$\frac{d\theta}{\theta} = -\frac{M F_1}{V_1 \gamma_1 c_1'} \left(\tau - \tau_0\right)^{n-1} d(\tau - \tau_0), \qquad (21)$$

since $d\tau = d \left(\tau - \tau_0\right)$.

Integrating, we finally obtain

$$\ln\theta = -\frac{MF_1}{nV_1\gamma_1 c_1'}(\tau-\tau_0)^n + C. \tag{22}$$

The integration constant C will be found from the initial condition at $\tau=\tau_1$

$$\theta = \theta_{1in} - \theta_{2in} = \theta_{in}.$$

where $\tau_1 =$ duration of the first stage of freezing, or the pouring time;
$\theta_{1in} =$ initial temperature of the casting at the moment the pouring is completed;
$\theta_{2in} =$ temperature counted off the initial temperature of the mould as from zero.

Substituting this condition into Eq. (22) we obtain

$$\ln\theta_{1in} = -\frac{MF_1}{nV_1\gamma_1 c_1'}(\tau_1-\tau_0)^n + C,$$

whence

$$C = \ln\theta_{1in} + \frac{MF_1}{nV_1\gamma_1 c_1'}(\tau_1-\tau_0)^n.$$

Eliminating the value of C from Eq. (22) we find that

$$\ln\frac{\theta}{\theta_{1in}} = \frac{MF_1}{nV_1\gamma_1 c_1'}[(\tau_1-\tau_0)^n - (\tau-\tau_0)^n]. \tag{23}$$

The time during which the casting is cooled to the temperature θ can be found from Eq. (24)

$$(\tau-\tau_0)^n = -\frac{n}{M}V_1\gamma_1 c_1' \ln\frac{\theta_{1in}}{\theta} + (\tau_1-\tau_0)^n h r^n. \tag{24}$$

The full duration of the first two stages of freezing τ (duration of the mould filling and the time during which the metal superheat is extracted) will be found from Eq. (24) replacing θ by the temperature difference $\theta_{1in} - \theta_{2in}$

$$(\tau_2-\tau_0)^n = \frac{n}{M}V_1\gamma_1 c_1' \ln\frac{\theta_{1in}}{\theta_{1in}-\theta_{2in}} + (\tau_1-\tau_0)^n h r^n. \tag{25}$$

The change in the temperature of the casting during the superheat transfer conforms to the equation

$$\frac{\theta}{\theta_{1in}} = e^{-\frac{M}{n}\frac{(\tau-\tau_0)^n-(\tau_1-\tau_0)^n}{V_1\gamma_1 c_1'}}. \tag{26}$$

Eq. (26) has been obtained from the solution of Eq. (23).

These equations describe the case of gradual extraction of heat from the casting produced in a sand or swabbed metal mould.

The solidification time of a casting from θ_l (liquidus temperature) to θ_s (solidus temperature) can be found in the same way by constructing a heat balance equation for the solidification period $\Delta\theta_{sol}=\theta_l-\theta_s$, like Eq. (22)

$$M(\tau-\tau_0)^{n-1}\theta_1 F_1\,d\tau = -V_1\gamma_1\left(c_{1m} + \frac{Q_1}{\Delta\theta_{sd}}\right)d\theta_1, \tag{27}$$

where $c_{1m} = \dfrac{c' + c_1}{2}$ is the mean specific heat capacity of the metal, equal to half the sum of the heat capacities of the metal in a solid and liquid state, kcal/kg °C;

Q_1 = heat of the metal solidification, kcal/kg.

Integrating Eq. (27) we obtain

$$\ln \theta_1 = -\frac{MF_1}{nV_1\gamma_1 c_{eff}}(\tau - \tau_0)^n + C, \qquad (28)$$

where $c_{eff} = c_{1m} + \dfrac{Q_1}{\Delta\theta_{sol}}$ is the effective heat capacity over the range of solidification temperatures, kcal/kg °C.

The integration constant is eliminated by substituting the initial conditions: at

$$\tau = \tau_2; \quad \theta_1 = \theta_1 - \theta_{2in} = \theta_{in}$$

we finally obtain

$$\ln \frac{\theta_1}{\theta_{in}} = \frac{MF_1}{nV_1\gamma_1 c_{eff}}[(\tau_2 - \tau_0)^n - (\tau - \tau_0)^n]. \qquad (29)$$

The time during which the casting solidifies to the temperature θ_1, which is within the range of the solidification temperatures $\Delta\theta_{sol}$, can be found from Eq. (29)

$$(\tau_1 - \tau_0)^n = \frac{n}{M}V_1\gamma_1 c_{eff}\ln\frac{\theta_{in}}{\theta_1} + (\tau_2 - \tau_0)^n \, hr^n. \qquad (30)$$

The time of full solidification of the casting τ_s, from the moment the mould begins to be filled to the end of solidification will be found from Eq. (30) by substituting in it the value θ_s for θ_1

$$(\tau_s - \tau_0)^n = \frac{n}{M}V_1\gamma_1 c_{eff}\ln\frac{\theta_l}{\theta_s} + (\tau_2 - \tau_0)^n \, hr^n. \qquad (31)$$

The temperature of the casting within the range of solidification temperatures is given by Eq. (29)

$$\frac{\theta_s}{\theta_l} = e^{\dfrac{\dfrac{M}{n}(\tau - \tau_0)^n - (\tau_2 - \tau_0)^n}{V_1\gamma_1 c_{eff}}} \qquad (32)$$

The thickness of the solidified skin of a flat casting is determined in conformity with the square root law, i. e., by Eq. (16)

$$\xi = k\sqrt{\tau}.$$

If the metal has filled the mould instantaneously, i. e., $\tau_1 = 0$, and if it has no superheat, i. e., $\tau_2 = 0$, the thickness of the solidified layer can be found from the following equation

$$\xi = 2M\frac{\theta_{sol}}{\gamma_1 Q_1}\sqrt{\tau}\,\mathrm{m}. \qquad (33)$$

The volume of the solidified metal of a flat plate will come to

$$V_1 = F_1\xi \, \text{cu m}, \qquad (34)$$

where F_1 = the freezing surface of the casting, sq m.

The volume of the solidified metal of a cylindrical casting cooled from outside (Fig. 23a) can be found geometrically

$$V = \pi x_1^2 - \pi(x_1 - \xi)^2 L \, \text{cu m}, \qquad (35)$$

where $L = $ length of the casting, m;

$x = $ radius of the interface between the casting and mould, m.

Transforming Eq. (35) we shall obtain

$$V = F_1 \xi \left(1 - \frac{1}{2} \delta \right) \text{cu m,} \qquad (36)$$

where $F_1 = 2\pi x_1 L$ is the freezing surface, sq m;

$\sigma = \dfrac{\xi}{x_1}$ is the relative thickness of the solidified skin, m.
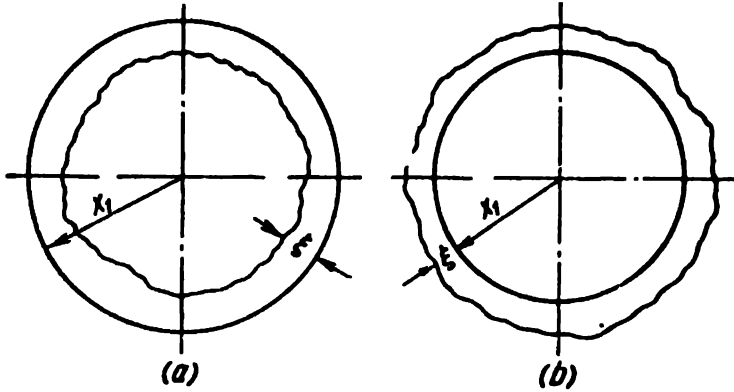
Fig. 23. Diagram of solidification of a cylindrical casting:

a—casting cooled from outside; *b*—casting cooled from inside

The same method is used to find the volume of the solidified skin of a spherical casting:

$$V = \frac{4}{3} \pi x_1^3 - \frac{4}{3} \pi (x_1 - \xi)^3,$$

whence

$$V = F_1 \xi \left(1 - \delta + \frac{1}{3} \delta^2 \right) \text{cu m,} \qquad (37)$$

where $F_1 = 4 \pi x_1^2$, sq m

$$\delta = \frac{\xi}{x_1}$$

The thickness of the solidified skin ξ is not proportional to the volume of a cylinder or sphere and as V increases the growth is speeded up, whereas the growth of the skin thickness of the plate remains all the time proportional to the volume [compare Eqs (36) and (37) with Eq. (27)].

The volume of the solidified skin of a hollow cylindrical casting (Fig. 23*b*) cooled from inside can be also found geometrically

$$V = \pi (x_1 + \xi)^2 L - \pi x_1^2 L,$$

whence

$$V = F_1 \xi \left(1 + \frac{1}{2} \delta \right) \text{cu m,} \qquad (38)$$

where $F_1 = 2\pi x_1 L$, sq m;

$$\sigma = \frac{\xi}{x_1}, \, m.$$

Here an increase in V slows down the growth ξ.

For a sphere which solidifies from inside we shall obtain, respectively:

$$V = \frac{4}{3} \pi (x_1 + \xi)^3 - \frac{4}{3} \pi x_1^3 , \qquad (39)$$

whence

$$V = F_1 \xi \left(1 + \delta + \frac{1}{3} \delta^2 \right) \text{cu m,}$$

where $F_1 = 4\pi x_1^2$, sq m;

$$\delta = \frac{\xi}{x_1} \text{m.}$$

When we compare the dependences obtained for the bodies of the simplest shape (plates, cylinders, spheres) we cannot fail to notice their unidimensional character. All the design equations are analogous to Eq. (34) for a flat plate and differ only in the multipliers in the right-hand side. They can be therefore represented by the following equation:

$$V = F_1 \xi \Gamma \qquad (40)$$

or

$$\frac{V}{F_1 \xi} = \Gamma, \qquad (41)$$

where Γ is the value which depends only on the shape of the casting, i. e., it is a geometrical factor.

Obviously, for a flat casting $\Gamma = 1$, for a cylinder cooled from outside $\Gamma = 1 - \frac{1}{2}\delta$, and from inside $\Gamma = 1 + \frac{1}{2}\delta$. For a sphere cooled from outside $\Gamma = 1 - \delta + \frac{1}{3}\delta^2$, and from inside $\Gamma = 1 + \delta + \frac{1}{3}\delta^2$.

The diagram in Fig. 24 and Table 6 have been compiled for the purpose of facilitating the calculations by the above formulas.

The geometrical factor Γ can be regarded as a measure of dissimilarity between castings of various shape brought about by the deviations in the thermal properties of castings or, in other words, by the rate of skin crystallisation. As the thickness of the solidifying skin changes in time, the thermal properties of castings of any shape approach the properties of flat castings of a plate type. But as the skin increases in thickness, the dissimilarity between the castings becomes more and more pronounced, especially when

$$\delta = \frac{\xi}{x_1} \longrightarrow 1.$$

These effects will be understood more readily if we compare the thicknesses of the solidifying skin

$$r = \frac{V}{F_1} = \xi \text{m.} \qquad (42)$$

The value r can be found by solving Eq. (34).

Table 6

Values of the Geometrical Factor Γ
(Veinik)

Relative thickness of solidified skin $\delta = \dfrac{\xi}{x_1}$	Cylinder		Sphere	
	cooled from outside	cooled from inside	cooled from outside	cooled from inside
0	1.00	1.00	1.00	1.00
0.1	0.95	1.05	0.90	1.10
0.2	0.90	1.10	0.81	1.21
0.3	0.85	1.15	0.73	1.33
0.4	0.80	1.20	0.65	1.45
0.5	0.75	1.25	0.58	1.58
0.6	0.70	1.30	0.52	1.72
0.7	0.65	1.35	0.46	1.86
0.8	0.60	1.40	0.41	2.01
0.9	0.55	1.45	0.37	2.17
1.0	0.50	1.50	0.33	2.33
1.2	—	1.60	—	2.68
1.4	—	1.70	—	3.05
1.6	—	1.80	—	3.45
1.8	—	1.90	—	3.88
2.0	—	2.00	—	4.33

For flat castings the values r and ξ are identical. For castings of any other shape the relationship between the values r and ξ can be established in a general form from Eq. (40)

$$r = \frac{v}{F_1} = \xi\Gamma, \qquad (43)$$

I. e., the reduced thickness of the solidifying skin becomes proportional to the geometrical thickness.

Fig. 24 shows that for cylinders and spheres cooled from outside $\Gamma < 1$ and for hollow castings $\Gamma > 1$.

In the former the skin grows at a fast rate and in the latter—slowly.

These considerations are the simplest proof of the fact that the value of the coefficient k in Eq. (16) is not constant and that the Stefan's law can be used only to calculate the solidification of flat castings.

In foundry practice castings of simple shape (plates, cylinders and spheres) are not very frequent. This can be attributed to the specific features inherent in foundry practice which usually has

to deal with complex shapes, as distinct from simpler forged and rolled products.

Complex shapes can be due to the presence of hollows, projections, joints, fins and holes in the casting. It goes without saying that the thickness of the skin in various portions of such a casting will increase differently. At first, when the solidification process extends only to insignificant volumes of the casting and mould, the profile of the growing skin will repeat the profile of the casting and its thickness will remain approximately uniform. As the process of solidification goes on, the crystallised metal
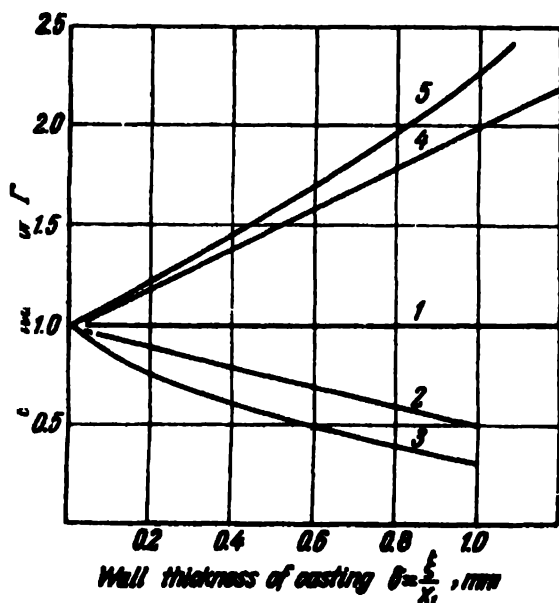
Fig. 24. Change in the geometrical factor Γ as a function of the wall thickness of the casting (Veinik):
I—flat casting; *2*—cylindrical casting cooled from outside; *3*—spherical casting cooled from outside; *4*—hollow cylindrical casting cooled from inside; *5*—hollow spherical casting cooled from inside

will grow in volume, the isothermal surfaces will move deeper inside and the irregularities of the surfaces being cooled will exert lesser effect on the shape of the interface separating the solid and liquid phases (solidification at a constant temperature) or on the shape of the two-phase zone (solidification over the temperature range).

Since the heat flux is stable in the depth of a heavy casting the isothermal surface will gradually approach a correct cylindrical or spherical configuration (Fig. 25). Hence, if the casting or its elementary area can be referred to one of the three investigated cases, the process of the skin growth can be studied with the aid of the above relationships for a flat plate, cylinder or sphere. When we plot the solidification isotherms on a drawing of a casting section (Figs 23 and 25) we must not let out of sight the effect of equalisation of the isothermal surface related to the action of the stability principle of the heat flux, which is sometimes referred to as angular effect.

The results of investigating the solidification process in variously shaped steel castings carried out by B. Gulyaev are represented in a graphical form in Fig. 26 as the coordinates—ratio

between the thickness x of the solidified skin of a casting and the referred thickness of the casting wall R versus the referred time of solidification $\frac{\tau}{R^2}$. The shape and size of the castings are illustrated in Table 7.

Table 7

Characteristics of Castings
(Gulyaev)

No. of casting	Shape	Half the thickness or the casting radius in cm
1	Plate	1.75; 2.30
2	Square bar	4.5
3	Sphere	5.62; 7.50; 11.20
4	Plate	0.5; 1.5; 2.5; 4.0; 5.0; 10; 17.5
5	Cylinder	7.5; 10; 20
6	Plate	0.7; 1.25; 2.00
7	Plate	2.5; 5.1
8	Square bar	6.35
9	Square bar	10.0
10	Cube	3.75

The symbols related to the solidification of the plates (castings Nos 1, 4, 6 and 7) are grouped around parabolas. The data obtained by Briggs and Gezelius correspond to the parabola in Eq. (16)

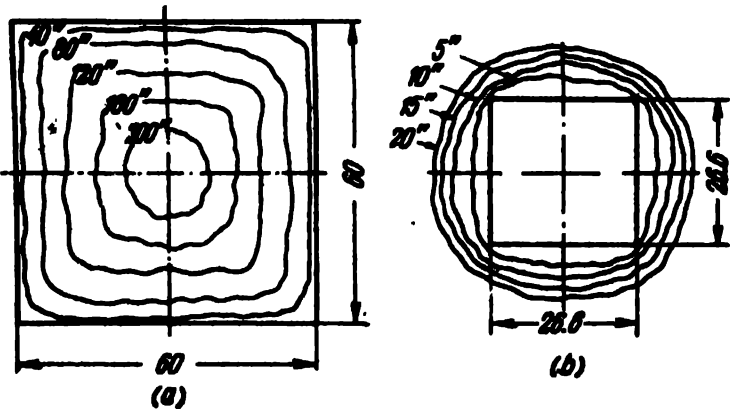*Fig. 25.* Solidification isotherms for a zinc casting (Veinik):

a—cooling from outside, the metal was poured after 40, 80, 120 and 200 sec; b—cooling from inside, the metal was poured after 5, 10, 20 and 30 sec

with $k=1.50$ cm/min$^{-1/2}$ (curve c), the Chvorinov data to the parabola with $k=0.59$ cm/min$^{-1/2}$ (curve a), while the symbols related to the other cases of plate solidification are plotted on

·curve *b* (*k*=1.20 cm/min$^{-1/2}$) (curve II). The symbols describing the spheres, cylinders and cubes (castings Nos 3, 5 and 10) are ·characterised by, an accelerated solidification of the axial portion ·of the casting (lower for the cylinder and higher for the sphere). The symbols which characterise the square bars are represented ·on the curve *d* and the symbols for the spheres and cube—on the ·curve *e*. The experimental values obtained by various investigators differ but slightly. The only exceptions are the data provided by N. Chvorinov for the plates; their deviations from the data ·amassed by all other researchers—Y. Nekhendzi, A. Levi and W. Ruff and the works mentioned above—are so large that one ·cannot but suppose that some systematic error has been made. The data provided by N. Chvorinov on the cylinders are in good ·agreement with the results obtained by other experimenters.

The time required for the complete solidification of the plates ·can be found from the following equation

$$\tau_1 = \left(\frac{R}{k}\right)^2.\tag{44}$$

The time actually found for them must coincide with the parabolas in Fig. 26. For cylindrical and spherical castings the ·solidification period will be reduced.

B. Gulyaev suggests that the solidification time for the elements of a shaped casting be determined from the equation

$$\tau_3 = M\Phi R^2,\tag{45}$$

where τ_3 = full solidification capacity, hr;

M = coefficient depending on the material and shape of the casting as indicated in Table 8, compiled by B. Gulyaev from the results of numerous investigations;

Φ = coefficient of shape, which is 1.00 for the plate, 0.76 for the cylinder and 0.47 for the sphere;

R = half of the thickness or radius of the element, m.

Consider several examples for the calculation of the solidification period.

1. Find the total time required by a flat casting 26 mm thick, to solidify in a sand mould. From Table 8, M=0.592 min/sq cm, for the plate ·Φ=1.00. Then, from Eq. (45), the solidification period will be

$$\tau_3 = 0.592 \times 1 \times 1.3^2 = 1 \text{ min.}$$

2. Find the time required by a steel shaft 1,000 mm in diameter to solidify in an iron-cast mould.

From Table 8, M=0.148 min/sq cm. For a cylinder Φ=1.76. From Eq. (45) we have

$$\tau_3 = 0.148 \times 0.76 \times 50^2 = 281 \text{ min.}$$
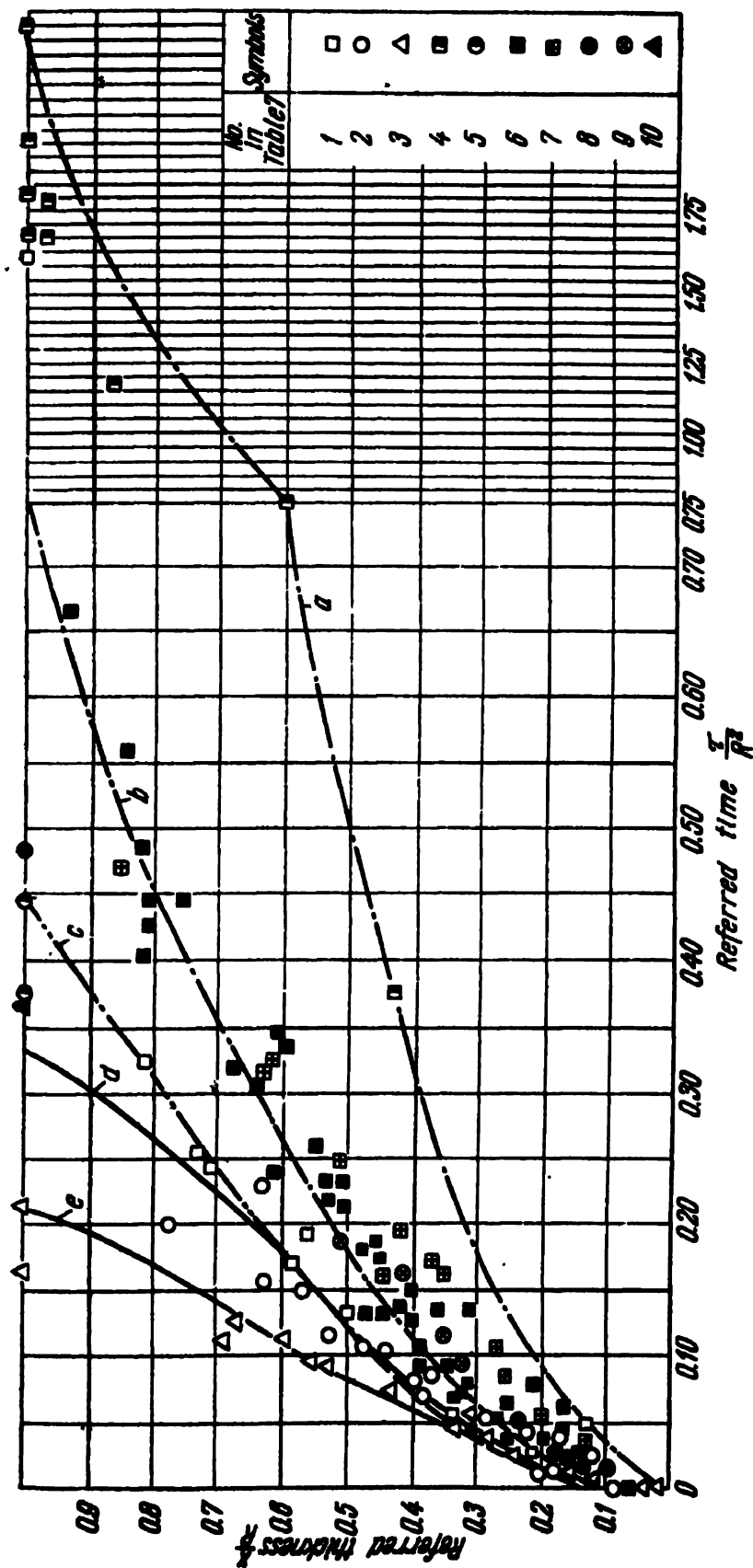
Fig. 26. Change in the referred time of steel casting solidification in a sand mould as a function of the referred thickness (Gulyaev):

a—from Eq. (16), at $k=0.69$ cm/min$^{\frac{1}{2}}$ for the plate; b—from Eq. (16), at $k=1.20$ c m/min$^{\frac{1}{2}}$ for the plate; c—from Eq. (16), at $k=1.50$ cm/min$^{\frac{1}{2}}$ for the plate; d—experimental data for the cylinder; e—experimental data for the sphere

No. in Table 7	Symbols	
1	□	○
2	◁	
3	◼	
4	◕	
5	◼	
6	◙	
7	●	
8	◉	
9		
10	▲	

Referred time $\frac{\tau}{R^2}$

Referred thickness ξ

Table 8

Values of the Coefficients *k* and *M*
(Gulyaev)

Material of casting	Freezing conditions	Solidification coefficient k, cm/min^{-1}/s	Coefficient of material M, min/cm²
Stearine	In air, tin mould	0.164	37.2
Stearine	In running water	0.245	16.7
Steel	In sand mould	1.3	0.592
Steel	In iron-cast mould	2.6	0.148
Pig iron:			
grey	In sand mould	0.7	2.04
grey	In metal mould	2.2	0.21
malleable	In sand mould	1.1	0.82
malleable	In metal mould	2.0	0.25
Brass	In sand mould	1.8	0.31
Brass	In iron-cast mould	3.8	0.07
Brass	In copper mould cooled with water	4.2	0.06
Aluminium	In iron-cast mould	3.1	0.10

12. SHRINKAGE, AFTEREFFECTS AND PREVENTION

Cast steel may develop a number of defects, most of which being caused by volumetric shrinkage or contraction during solidification. If it were not for contraction in steel such defects as shrinkage cavities and porosity would be altogether impossible, while hot and cold cracks would be extremely rare.

Back in 1878, D. Chernov investigated the principal causes of steel defects and evolved the methods for their remedy. He wrote: *

"The main shortcoming inherent in steel castings is the presence of voids and shrinkage cavities and sometimes of cracks at the surface.... Despite these shortcomings we cannot deny the obvious advantages that accrue from the manufacture of products by casting in moulds. Since only a thorough knowledge of any phenomenon will provide effective means for its prevention, it will be easily understood why much time and effort are given at modern iron and steel works to gain an insight into the causes of defects in cast metal."

These words did not lose their meaning in our day, too.

* D. Chernov. *Science of Metals*, Metallurgizdat, 1950.

(a) Amount of Shrinkage

As the steel poured into a foundry mould cools down, the atoms draw nearer to one another and contract the initial volume of the steel. This phenomenon is called shrinkage or contraction.

There are liquid shrinkage, on cooling in the liquid state (ΔV_l), and solidification shrinkage, on transforming from the liquid to the solid state (ΔV_{sol}), which in their physical sense can be called volumetric shrinkages, as well as solid shrinkage, on cooling in the solid state (ΔV_s), which can be regarded as linear.

Shrinkage is usually understood to mean a percentage change in volume or length, i. e.,

$$\Delta V = \frac{V_0 - V_1}{V_1} \, 100 \text{ per cent};$$

$$\Delta l = \frac{l_0 - l_1}{l_1} \, 100 \text{ per cent};$$

where V_0, V_1 = initial and final volume, cu cm;
l_0, l_1 = initial and final length, cm.

A free linear shrinkage of a casting made from steel 35Л comes to 2.5 per cent, from alloy structural steel 35ХЛ to 2.44 per cent, from steel 35ХМЛ to 2.48 per cent, from steel 30ХНВЛ to 2.42 per cent, and from steel 30ДСХНЛ to 2.38 per cent.

In essence, linear shrinkage is a shrinkage on cooling in the solid state. For this reason the volumetric shrinkage in the solid state is three times as large as the linear shrinkage and amounts to 7.5 per cent for steel 35Л.

The change in the volume of metal in a casting is not identical to the change in the volume of the casting itself due to the evolution of gases and allotropic transformations, as well as the deformation of the mould.

The full volumetric shrinkage of metal can be found from the equation

$$\Delta V = \Delta V_l + \Delta V_{sol} + \Delta V_s. \tag{46}$$

The addends of this sum can be represented in the form of the following equations:

$$\Delta V_l = \alpha_l \, (\theta_f - \theta_l); \tag{47}$$

$$\Delta V_{sol} = \alpha_{sol} \, (\theta_l - \theta_s);$$

$$\Delta V_s = \alpha_s \, (\theta_s - \theta_0), \tag{48}$$

where $\dot{\alpha}_l$, α_{sol}, α_s = coefficients of liquid, solidification and solid shrinkage, respectively;

θ_f, θ_l, θ_s, θ_0 = temperature during filling the mould, the liquidus and solidus temperatures and the temperature of a casting being cooled at the given moment, respectively.

Separate stages in the volumetric shrinkage of steel are characterised by different changes in specific volume (Fig. 27). After a sharp drop of temperature during the contraction of liquid steel and during its solidification the specific volume curves change from an almost vertical position to an inclined one. From the
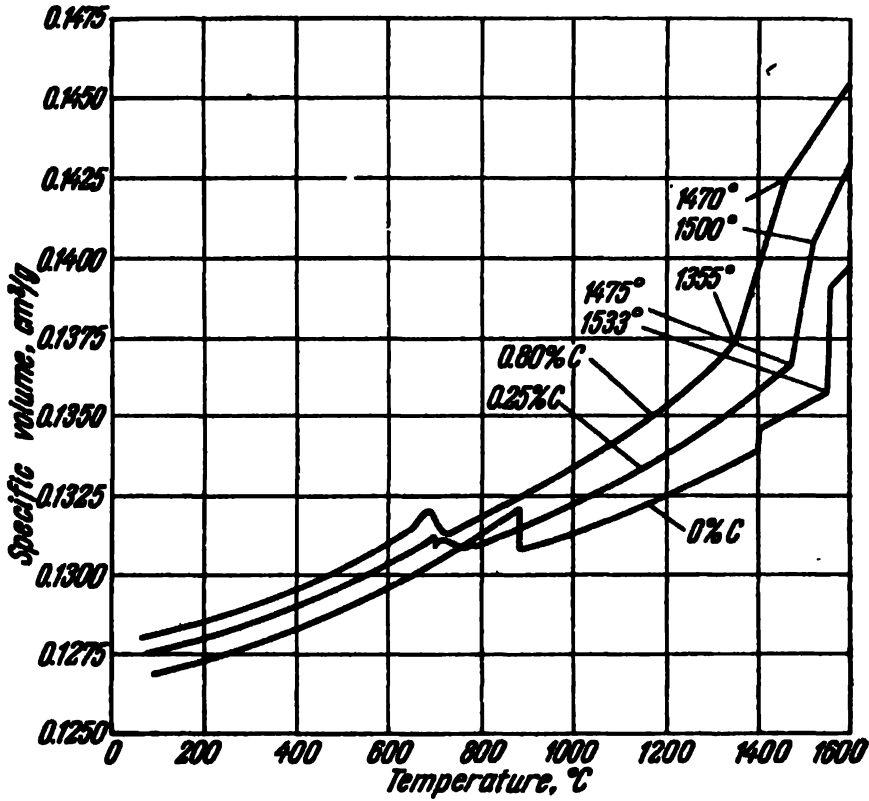
Fig. 27. Change in the specific volume of carbon steel as a function of carbon content and temperature

solidus temperature to the temperature of pearlitic transformations there occurs contraction of γ-iron, called the pre-pearlitic shrinkage (α_{pre}). Then, during the $\gamma \to \alpha$ transformation there takes place the phase expansion ΔV_{ph}, and further, from the temperature of the transformation and to room temperature, we observe the contraction of α-iron known as the post-pearlitic shrinkage (α_{post}).

In connection with the post-pearlitic shrinkage, the term ΔV_s in Eq. (46) can be represented in the following form:

$$\alpha V_s = \alpha_{pre}(\theta_s - \theta_{\gamma \to \alpha}) \Delta V_{ph} + \alpha_{post}(\theta_{\gamma \to \alpha} - \theta_0). \qquad (49)$$

As the superheat temperature of liquid steel rises, its contraction grows in scope [see Eq. (47)]. A broader solidification temperature range likewise increases the steel contraction on freezing. This explains the growth of the full volumetric shrinkage at a

higher carbon content in steel: at 0.10, 0.40, 0.70 and 1.00 per cent C the volumetric shrinkage amounts to 10.5, 11.3, 12.1 and 14.0 per cent, respectively.

Various alloying elements affect differently the specific volume and, consequently, the shrinkage of steel (Fig. 28).
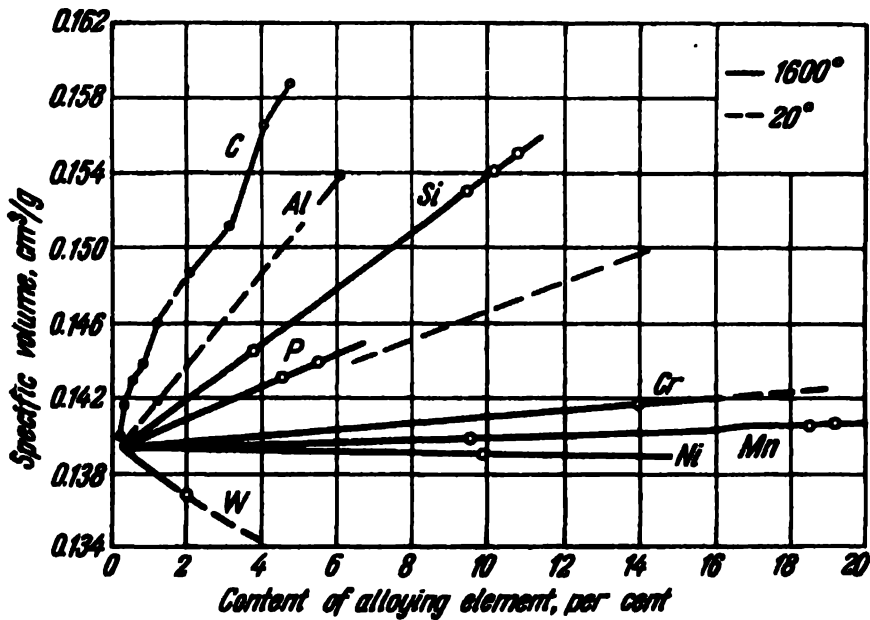
Fig. 28. Change in the specific volume of liquid steel at 1600° C and of solid steel at 20° C as a function of the alloying element (from data obtained by various investigators)

(b) Characteristics of Shrinkage Cavities

A shrinkage cavity is a void in the casting material usually of a conical or spindle-like shape with a rough coarse-grained surface. The roughness owes its origin to the different development of the dendrites and is the principal feature by which shrinkage cavities can be distinguished from gas holes.

Shrinkage cavities are the direct consequence of a much higher coefficient of contraction in liquid steel than is observed in solid steel.

If a cube with the volume V_1 is instantaneously filled with metal there will be no hard skin formed on the mould-metal interface and the temperature of the metal will be the same at any point of the volume. Heat exchange will start directly after pouring: the heat lost by the steel will be absorbed by the mould. Every elementary area of the cube will be cooled at a different rate: faster outside and slower inside. A temperature difference will arise between the centre of the cube and its surface, hottest in the centre. The specific volume of steel will be obviously greater where the temperature is highest. Hence, the temperature differ-

ence will produce a difference in the specific volumes. It is likewise obvious that in equal volumes the number of the molecules of iron and its admixtures will increase in inverse proportion to the specific volume of the metal. A decrease in the volume of the colder metal will cause a local vacuum to appear and, as a result, a pressure difference in the neighbouring portions of the casting. The liquid metal will be moved by the gravity force and the sucking action of the colder metal from hotter towards colder areas, mainly from top to bottom, and from the centre towards the periphery of the casting.

After a lapse of certain time the temperature of the metal on the cube faces will drop to the solidification temperature and a thin skin of solid metal will form on the mould-metal interface. The volume of the cube will decrease to V_2. Only the volumetric shrinkage of liquid metal will take place during this time interval, as a result of which its level in the mould will decrease and the cube will become a parallelepiped. There will arise a difference between the volumes of the mould cavity and the solidifying casting, i. e., a void will develop between the mould and the casting, forming what might be called an "external" shrinkage cavity. So far, no shrinkage cavity has developed inside the casting.

Some time later the skin will grow in thickness and it will be able to resist the pressure of the liquid metal and to preserve the shape of the casting without the aid of the mould. Subsequently, the elementary volumes of the liquid metal will continue to move but the process of metal shaping will undergo essential changes—the casting will decrease not only in height but also in all its other dimensions. When there is much liquid metal inside the casting its volume will decrease during the same period much more than the external dimensions of the casting with a lesser specific volume of the solid metal. Hence, the pronounced difference between the coefficients of the volumetric shrinkage in the liquid and solid metal will cause a break between the casting and the mould and a shrinkage cavity will develop in the upper, hotter part of the casting (Fig. 29).

Towards the end of solidification of the entire casting the size of the shrinkage cavity will grow to the largest extent possible and its development will cease. And although in subsequent freezing the absolute volume of the shrinkage cavity will somewhat decrease, its volume, as related to the volume of the casting, will remain unaltered.

In this way, a shrinkage cavity may develop only during solidification.

According to Y. Nekhendzi and I. Girshovich, the relative volume of a shrinkage cavity is

$$\Delta V_{cav} = \alpha_l \left(\theta_{l.m} - \theta_{sol}\right) + \Delta V_{sol} - 1.5\alpha_s \left(\theta_{sol} - \theta_{s.m}\right), \quad (50)$$

where a_l = coefficient of linear shrinkage of liquid metal at a corresponding temperature;

$\theta_{l\cdot m}$ = mean temperature of liquid metal inside the casting at the start of solidification, °C;

θ_{sol} = solidification temperature, °C;

ΔV_{sol} = volumetric shrinkage in a state of solidification;

$a_.$ = coefficient of linear shrinkage of solid metal at a corresponding temperature;

$\theta_{s\,m}$ = mean temperature of solid metal at the end of solidification of the entire casting, °C.

Fig. 29. Concentrated strinkage cavity in a steel casting

According to various investigators, the volume of the shrinkage cavity amounts to 6-8 per cent of the casting volume.

It follows from Eq. (50) that the volume of a shrinkage cavity is dependent primarily on the contraction of the liquid metal inside a solidifying casting and the contraction during solidification.

The size of shrinkage cavities can be reduced by:

(a) decreasing the coefficient of the volumetric shrinkage of liquid metal;

(b) decreasing the temperature of steel during pouring;

(c) decreasing the pouring rate;

(d) increasing the heat conductivity of metal;

(e) decreasing the thickness of the casting;

(f) decreasing the temperature range of solidification;

(g) decreasing the coefficient of the volumetric shrinkage during solidification;

(h) building up a relatively thick hard skin in pouring;

(i) taking all measures tending to equalise the temperature across the casting section;

(j) increasing the shrinkage coefficient of solid metal;

(k) decreasing the mean temperature of solid metal towards the end of solidification.

As a rule, a shrinkage cavity develops in the form of a parabolical cone. This can be observed in a casting with the height H and radius R cooled only from the sides (Fig. 30).
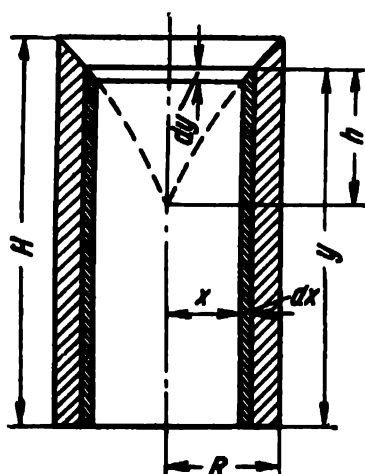
Fig. 30. Calculation of the mould and the location of a shrinkage cavity in a cylindrical casting cooled only from the sides (Nekhendzi):

R—radius of the casting; H—height of the casting; h—depth of the shrinkage cavity; x and y—variable coordinates of the solidifying layer
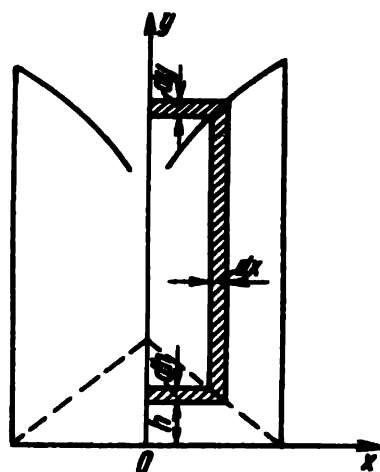
Fig. 31. Shrinkage cavity in a casting cooled from below and from the sides (the same symbols as in Fig. 30)

Let us assume that a hard skin, x thick, has formed in the casting during the time τ. After a lapse of time $d\tau$, there appeared another layer, dx thick, with the height y. The shrinkage of this layer will be compensated for by liquid metal with the volume $\pi x^2 dy$.

Hence,

$$2\pi\, xydx\, \Delta V_{sol} = \pi x^2\, dy.$$

To simplify our case we shall consider only solidification shrinkage.

After appropriate transformations and integration we shall
have

$$2\Delta V_{sol} \ln x = \ln y + C. \qquad (51)$$

After finding the integration constant C from the initial conditions, when at $x=R$, $y=H$, we shall obtain finally

$$y = H\left(\frac{x}{R_s}\right)^{2\Delta V_{sol}} \qquad (52)$$

The same method can be used to develop the equation for a shrinkage cavity in a casting cooled from the sides and from below (Fig. 31).

It follows from Eq. (52) that the depth of penetration of a shrinkage cavity is limited only by the point where the bottom and side solidification fronts merge.
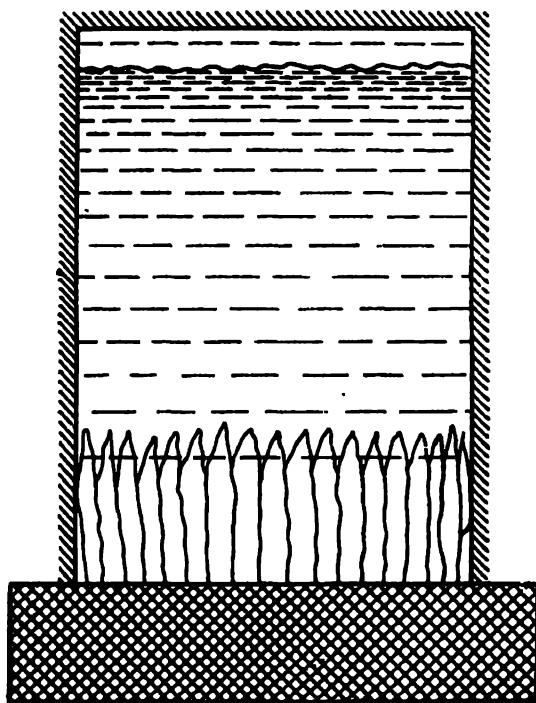
Fig. 32. Solidification of a casting cooled only from below

Let us consider several typical cases of solidification in a mould after it has been filled with liquid metal.

Let us assume that the heat is extracted only from the bottom of the mould, for example with the aid of a large external chill. There are no heat losses from the top and the sides (Fig. 32). The layer of metal in contact with the chill will quickly solidify and separate from the sides of the mould.

In this case there will be no temperature difference across the section of the casting. Since the metal solidifies gradually from

bottom to top, the level of the liquid metal will consistently decrease until it completely solidifies and no shrinkage cavity will be formed. Fins caused by the metal entering the parting lines between flasks may form on the vertical surfaces of the casting.

If during the intensive cooling from the bottom as before, heat is slowly extracted from the side surfaces and the top, the shrinkage cavity will take a lenticular, rather than conical, form (Fig. 33).
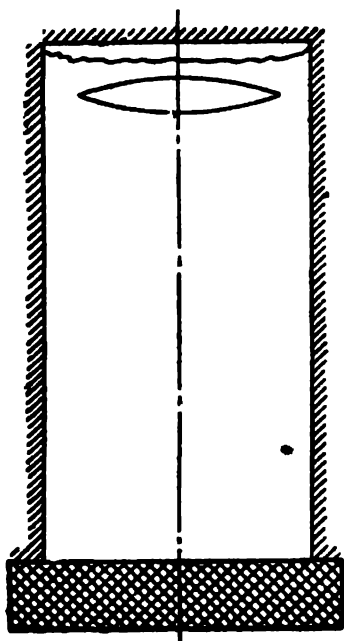
Fig. 33. Lenticular shrinkage cavity

Both cases can be observed in practice when the casting is heated from the top by various methods. Cooling of the casting from below will speed up the solidification process and from above retard it. In either case, there will be a considerable temperature difference between the upper and lower parts of the casting. An intensive heating from above, for example with a voltaic arc or induction heating, will ordinarily prevent the formation of shrinkage cavities.

These conclusions are in good agreement with Eq. (52). Indeed, when cooling from below or heating from above, the point where the advancing fronts will merge will be located at a considerable distance from the bottom of the casting (due to a more intensive solidification of the bottom and weaker solidification on the sides). In the limit, this may occur in one plane, i. e., there will be no shrinkage cavity.

Practical experience shows that the higher this point, the smaller will be the shrinkage cavity. Thus, when the risers are

heated with exothermal mixtures, the shrinkage cavity becomes smaller.

Under ordinary production conditions the castings are cooled from all sides at approximately the same rate. In this case, and due to the cooling effected from above, the shrinkage cavity will penetrate deeper into the casting. If, for some reason or other, no hard skin has been formed in the upper part of the casting, the shrinkage cavity will come to the surface.

(c) Effect of the Solidification Temperature Range on the Nature of Shrinkage Cavities

The process of the development of shrinkage cavities was considered above from the standpoint of heat engineering. Solidification and, particularly, the solidification temperature range are also extremely important in this process. A. Bochvar and A. Spassky were the first to study this problem and lay the foundations for the theory of solidification and contraction of alloys.
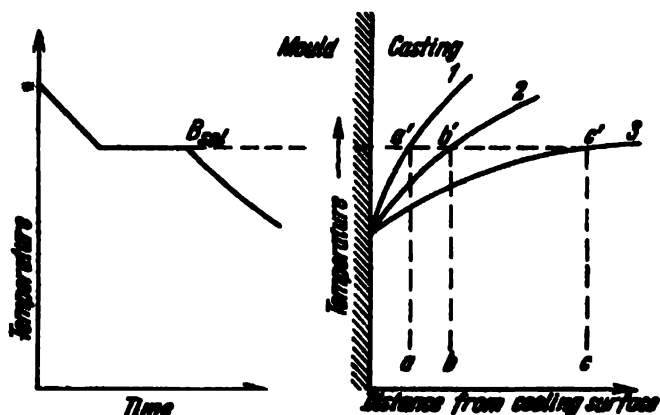
Fig. 34. Regions of metal solidification along the vertical section of a casting made from an alloy solidifying at a constant temperature (Spassky)

Let us first consider the solidification of a metal or alloy of a eutectic composition. The cooling curve (Fig. 34, left) shows an interruption of temperature which corresponds to the identical temperature of solidification (a pure metal or alloy of eutectic composition solidifies at a constant temperature). Curve *1* of the temperature difference, which corresponds to the first moment of time under study after the mould has been filled, shows that to the left from the dash line *aa'* (Fig. 34, right) part of the casting adjoining the mould surface has solidified and that to the right from this line the alloy is still in a liquid state since the temperature at any point of this section of curve *1* is higher than θ_{sol}. At the next instant of time (curve *2*) the solidification front will reach the boundary *bb'*. At this moment the boundary of the solid crystallites will progress towards the vertical line *bb'*. Consequently, to the left from this vertical line the temperature will be be-

low θ_{sol} at all the points of the volume and the alloy will be in a solid state, while to the right from the line bb' the temperature at all the points of the casting volume will be above θ_{sol} and the alloy will be in a liquid state.

Finally, at some later instant of time (curve 3) the solidification front will move to the right and reach the vertical line cc'. To the right from this vertical line the temperature in the entire zone will be maintained at θ_{sol}, i. e., it remains constant till the entire volume of the casting solidifies.

Let us note that at the first or second instants any solid crystallite brought into the zone located to the right from the line bb' or, respectively, from the line cc', will melt since the temperature here is above θ_{sol}. Conversely, if a drop of liquid penetrates into the zone to the left from the line aa' it will be immediately crystallised. During the first and second instants of time columnar crystallites grow in size. The casting to the left from the boundaries aa' and bb' must be dense in structure, without shrinkage microcavities. Each columnar crystallite grows unhindered being fed with the liquid in direct contact with the front of solid crystallites.

In its temperature distribution inside the casting the third instant differs in principle from the second one in that the temperature remains constant to the right from the boundary cc' since curve 3 becomes horizontal.

Practical experience knows of cases when the liquid entrained with it solid skeletons of dendrites detached during the stream movement. Such hard crystallite does not melt in the zone to the right from the line cc'. But neither can it develop, as long as the temperature remains constant. As the heat is being gradually removed, the temperature in the zone to the right from the line cc' begins to decrease after the third instant of time. Liquid will begin to precipitate on the surface of the entrained crystallite and convert to a solid state. The liquid molecules will agglutinate over the entire surface of the crystallite because there are no possibility for the predominant agglutination of ever new molecules on the top of the column due to diffusion, as is the case during the first and second instants. The crystallites will grow in all directions and assume an equiaxed, rather than columnar, form. In this way, when the temperature is distributed as shown by curve 3, the solidification conditions will favour the formation of equiaxed crystals and hamper the appearance of columnar crystals.

The temperature range solidification of any hypoeutectic or hypereutectic alloy (Fig. 35) differs from the scheme considered above. Let us take any binary alloy: m per cent of the component B, which forms solid solutions. Let us consider four interconnected diagrams (Fig. 35a-d).

At the moment *I* (after the mould has been filled) the temperature distribution in the casting can be represented graphically by the curve *I-I'*. The temperature is higher in the centre of the casting than at the surface which gives off heat to the cold walls of the mould (diagram *b*). The volume of the casting will be maximum (point *I* on diagram *c*).

At the moment *II* (start of solidification on the casting surface) the temperature distribution will be described by the curve *II-II'* and the volume by the point *II* (diagram *c*).
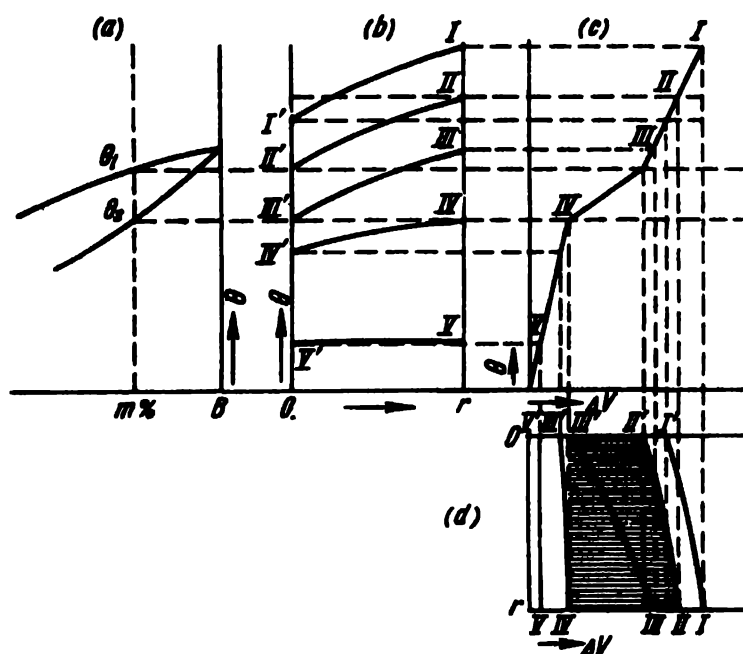
Fig. 35. Determining the size of a shrinkage cavity in a casting made of an alloy solidifying over the temperature range (Spassky):

a—constitutional diagram; *b*—diagram of temperature difference in the casting sections; *c*—diagram of the change in the volume ΔV of metal during contraction as a function of temperature θ; *d*—diagram of the change in the volume ΔV of a casting depending on the distance *r* from the surface

At the moment *III* (end of solidification of the alloy on the surface) the temperature distribution inside the casting can be represented graphically by the curve *III-III'* and the volume by the point *III* (diagram *c*).

At the moment *IV* (end of solidification of the centre-line zone of the casting) the temperature distribution is characterised by the curve *IV-IV'* and the volume by the point *IV* (diagram *c*).

At the moment *V* the casting will be cooled to the temperature of the environment which is the same in all the points of the volume (horizontal line *V-V'*) and the volume will be described by the point *V* (diagram *b*).

Depending on the distance from the surface, the casting shrinkage (diagram *d*) at the moment *I* will be represented by the

curve *I-I'*. But since the alloy is liquid in all the points of the volume, the level of the liquid in the casting and the value ΔV will decrease.

Solidification will commence on the surface at the moment *II*. The temperature *II'* corresponds to the liquidus temperature (point θ_l on the diagram). By transferring this point from the diagram *c* onto the diagram *d* we shall find the position of the point *II'* on the casting surface, which indicates the volume ΔV. The curve of the volume distribution across the casting section on diagram *d* will be expressed by the curve *II-II'*. The level of liquid in the casting will again go down, but its surface will develop a skeleton of solid crystallites, at first in the form of a lattice which indicates the position of point *II'* of the solidifying volume (diagrams *c* and *d*).

By the time described by the moment *III* the solidification on the surface of the casting will come to an end while in the rest of the volume it will proceed in conformity with the temperature distribution along the curve *III-III'*. The change in the metal volume across the casting section will be expressed on the diagram *D* by the area between the curves *II-II'* and *III-III'*.

A. Bochvar believes that the external dimensions of a casting are determined from the moment of the formation of a sufficiently rigid crystal skeleton capable of sustaining the liquid pressure and retaining the shape of the casting without the aid of the mould. In this way, the casting takes its shape within the temperature range θ_l and θ_s, i. e., between the curves *II-II'* and *III-III'* (diagrams *b*, *c* and *d*). That is, when there occurs a further decrease in the liquid level inside the area limited by the external hard skin, i. e., when a shrinkage cavity begins to form.

During the time interval between the curves *III-III'* and *IV-IV'* (diagram *b*) the moment of the end of solidification will move from the surface towards the centre of the casting along the dash horizontal line *III'-IV*. The changes in the metal volume during this period are characterised by the difference in the abscissas between the corresponding points on the curves *III-III'* and *IV-IV'* (diagram *d*). During this period the volume of metal will contract in two ways: due to the formation of solid crystallites in the liquid alloy and due to the reduction of the size of the previously solidified layer of metal. Shrinkage voids will appear between the pores of the crystallites over a considerable volume of the casting. The contraction of the volume of the solid layers of metal on diagram *d* corresponds to the distances between the vertical line *III'-IV* and the curve *IV'-IV*.

The volume of the casting and its dimensions, as well as the volume of the shrinkage voids, will decrease over the temperature range between the curves *IV-IV'* and *V-V'*.

Proceeding from the description of the mechanism governing

the contraction of the casting volume, and depending on the period between the end of pouring and complete cooling, diagram *d* can be used to find the area between the curves *I-I'* and *II-II'* which shows the contraction of the volume and the decrease in the alloy level. There are no shrinkage cavities as yet.

The area between the curve *II'-II* and the vertical line *II'* characterises the contraction of the volume of liquid metal after a solid layer has formed on the surface.

This volume begins to participate in the formation of shrinkage cavities. The area between the vertical lines *II'* and *III'* characterises a considerable reduction in the volume of metal during solidification. This volume likewise takes part in the formation of cavities.

Finally, the area between the vertical lines *III'-IV* and *V-V'* indicates the amount of contraction due to the solid metal being cooled. The size of the shrinkage cavities does not change relative to the size of the casting but decreases in its absolute value.

The full contraction of volume is determined by the area between the curve *I-I'* and the vertical line *V-V'*.

The size of the shrinkage voids depends on the position of the curve *II-II'* or, to be more precise, on the position of the point *II*—the temperature in the central layer of the casting when the surface reaches the temperature θ_t. If the mould is being filled with a superheated alloy the right end of the curve *II-II'* (diagram *b*) will go upwards and the lower end of the curve *II'-II* (diagram *d*) will shift to the right; the area bounded by the curves *III'-IV-II-II'* and the size of the shrinkage voids will increase.

A rapid filling of the mould will produce the same results. Slow pouring reduces the size of cavities, as shown by the temperature difference curves (diagram *b*).

Solidification of an alloy over the temperature range produces areas with a two-phase alloy, and the metal cannot fill completely the shrinkage microvolumes formed in the intercrystalline spaces, which results in shrinkage porosity.

A pure metal or alloy of a eutectic composition crystallises at a constant temperature. Of the solidification cases considered above, the most frequent are those which occur on the sides while the top solidifies more slowly, being sometimes so much retarded that its rate can be neglected. This gives us the simplest kind of solidification on all the sides and from the bottom (see Fig. 33).

The directional solidification, pressure difference and gravity forces combine to produce a concentrated funnel-shaped shrinkage cavity. The metal from this section is expended for the purpose of feeding the underlying contracted volume of the casting. The solidification of an alloy over the temperature range takes an altogether different course. Although the casting solidifies gradually, it becomes crystallised at once throughout the entire

volume. Each equiaxed crystallite contracts by itself due to shrinkage. At the end of the almost simultaneous completion of solidification the increase in the volume of the crystallite and its growth cease, because all of the liquid alloy has already solidified and gone to feed the contraction of the neighbouring crystallites. Separate microcavities, called shrinkage porosity, remain between the crystallites.

In actual fact, as the steel solidifies, part of the volume exhibits a columnar structure of the outer layer of a small ingot (Fig. 36), while the inner part consists of interlaced branches of equiaxed crystallites with spaces between them. Since the loose structure is concentrated in the axial zone it is called the centreline sponginess due to contraction.

There are three types of shrinkage porosity.

1. Centreline porosity which forms in the microscopic voids between the dendrites in the central zone of a casting when the flow of the feed liquid metal decreases and fails to compensate completely thin-walled and heavy castings. In pure metals and alloys of a eutectic composition the areas of centreline porosity
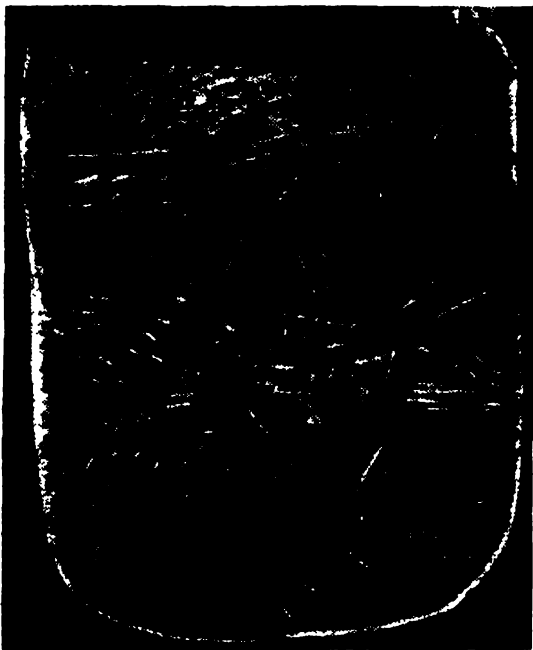
Fig. 36. Longitudinal section through a test casting. Shrinkage pores can be seen between the disconnected crystallites (Chvorinov)

are very well delineated, while in the alloys which solidify over the temperature range they extend to a fairly large volume of the casting with diffused boundaries.

2. Dispersed porosity which develops in the microscopic voids between the dendrites in conditions of free flow of liquid metal from the unsolidified part of the casting interrupted at a certain instant of time by the growing dendritic branches which block the flow of metal. Such porosity can be found in heavy

castings obtained in a sand mould at a low rate of alloy solidification and over a broad temperature range.

3. Localised shrinkage porosity which develops in separate parts of the casting, for example in thickened sections, isolated from the feeding sources. This porosity exhibits comparatively large voids whose size depends on the isolated volume of liquid metal.

The conditions which determine the start of sponginess grow more favourable when the rate of movement of an elementary volume of liquid metal in the centre of the casting $(\frac{dl}{d\tau})_{l=0}$, where it is connected to the feeding source, as well as the rate of metal flow governed by the laws of uninterrupted flow v_f, and the rate of liquid movement determined by shrinkage conditions with full compensation of all the shrinkage voids v_s, become identical, i.e.,

$$\frac{dl}{d\tau}\bigg)_{l=0} = v_f = v_s. \tag{53}$$

The elementary volume dV_v of the voids depends on the product $\Delta V_c'$ of the shrinkage cavity of relative volume by the elementary volume dV separated in the casting

$$dV_v = \Delta V_c' \, dV. \tag{54}$$

Since we are considering the time interval from the pouring of metal into the mould to the end of solidification, Eq. (52) shows that $\Delta V_c'$ is equal to the sum of the volumetric shrinkages of liquid steel on solidification.

Let us find the volume of the voids in the simplest case, i. e., for a rectangular plate L long and $2R$ thick, and so wide that the cooling of its end-faces can be neglected. Let us cut out of the plate a strip with a width equal to unity. Since this strip is fed with metal on the top of its end-face, a layer with a thickness of $x=R-r$ will solidify on the entire surface in time τ. When the interface moves by the distance dr, an element of liquid metal will be displaced by dl.

On the basis of Eq. (54), we can write

$$2L\Delta V_c' \, dr = 2rdl. \tag{55}$$

Expressing dr in terms of the time determined during freezing from the square root equation we shall obtain

$$\frac{dl}{d\tau} = \frac{\Delta V_c' \cdot kL}{2\sqrt{\tau}\,(R-k\sqrt{\tau})}. \tag{56}$$

Hence, the rate of liquid flow, which ensures the filling of all the voids, can be determined from the equation

$$v_v = \frac{\Delta V_c' k^2 L}{2(R-r)r}. \tag{57}$$

Since the intercrystalline space is too small, the flow of the feeding liquid will be laminar, affected by the force P, per unit of the section of the plate liquid portion. The entire section, with a width equal to unity and $2R$ thick, is acted upon by the force equal to $2\varrho P$. On the boundaries of the liquid portion of the plate, this force will be equalised by the friction force

$-2L\eta\dfrac{dv}{d\varrho}$, where η—dynamic viscosity and v—rate of liquid flow. Hence,

$$2\varrho P = -2L\eta\,\dfrac{dv}{d\varrho}$$

or

$$dv = -\dfrac{P_\rho d_\rho}{\eta L}\,. \tag{58}$$

Integrating within O-v and r-ϱ, we obtain

$$v = \dfrac{P(r^2 - \varrho^2)}{2\eta L}\,. \tag{59}$$

The consumption of the liquid to feed the voids can be found from the formula

$$V = \int_0^r vr d\varrho = \int_0^r \dfrac{P\,(r^2 - \varrho^2)}{\eta L}\,d\varrho = \dfrac{2}{3}\,\dfrac{Pr^3}{\eta L}\,. \tag{60}$$

The mean rate of liquid metal flow within the solidification boundaries will be

$$\dfrac{V}{2r} = V_b = \dfrac{Pr^2}{3\eta L}\,. \tag{61}$$

The force P is equal to the sum of the gas pressure in the shrinkage cavity and the pressure head γH, where H is the distance from the plate axis to the end of the main shrinkage cavity at the feeding source

The pressure of the gases in a shrinkage cavity which comes to the surface is equal to the atmospheric pressure, and in a cavity below the surface—to the pressure of the gases evolved from the steel on cooling. If the steel is not supersaturated with dissolved gases the gas pressure in the cavity below the surface may reach the vacuum level. If a gas producing charge is placed into the riser, the gas pressure may rise as high as 3-4 atm. And, finally, when molten steel is press-formed in a die with a hydraulic press the pressure may reach 500-700 atm.

In the ordinary case of a metal cast in a sand mould with the pressure in the riser of P_a, the total pressure P will become equal to the sum of the pressures in the riser and the pressure head:

$$P = P_a + \gamma H\,. \tag{62}$$

Equating Eqs (57) and (61) we shall obtain

$$\dfrac{\Delta V_c' k^2 L}{2\,(R - r)\,r} = \dfrac{(P_a + \gamma H)^2_r}{3\eta L}\,; \tag{63}$$

$$\dfrac{3a\eta k^2 L}{2\,(P_a + \gamma H)} = Rr^3 - r^4\,. \tag{64}$$

Since the value r^4 is small, it can be neglected, and then

$$r_s = \sqrt[3]{\dfrac{3\Delta V_c' \eta k^2 L^2}{2\,(P_a + \gamma H)\,R}}\,, \tag{65}$$

where r_s= half the thickness of the centreline sponginess, cm.

B. Gulyaev developed the equation for the radius of the centreline spongy zone of a prism or cylinder, which is very similar

Table 9

Design and Actual Centreline Sponginess in Steel Castings
(Gulyaev)

Size of casting, mm			Size of riser, mm		Sponginess, mm	
thickness	length	width	diameter	height	design	actual
10	1,000	100	100	45	1.03	
20	1,000	100	100	100	0.80	
20	1,000	100	100	550	0.72	
20	1,000	1,000	160	120	1.25	0.5-1.5
20	500	100	100	100	0.51	
40	1,000	100	100	100	0.64	

to the calculation of the plate solidification:

$$r_s = \sqrt[3]{\frac{32 \Delta V'_c \eta L^2 a^{3/2}}{3 (P_a + \gamma H) R^{1/4}}} .\tag{66}$$

The Eqs (65) and (66) have been verified in experiments on steel castings of various size. The results are illustrated in Table 9.

The following constant values have been accepted for the calculations: $\Delta V'_c = 0.085$; $\eta = 0.085$ g-cm/sec; $\gamma = 6.9$ g/cm^3; $P_a = 1,000$ g/cm^2; $K = 0.17$ cm/sec$^{\frac{1}{2}}$.
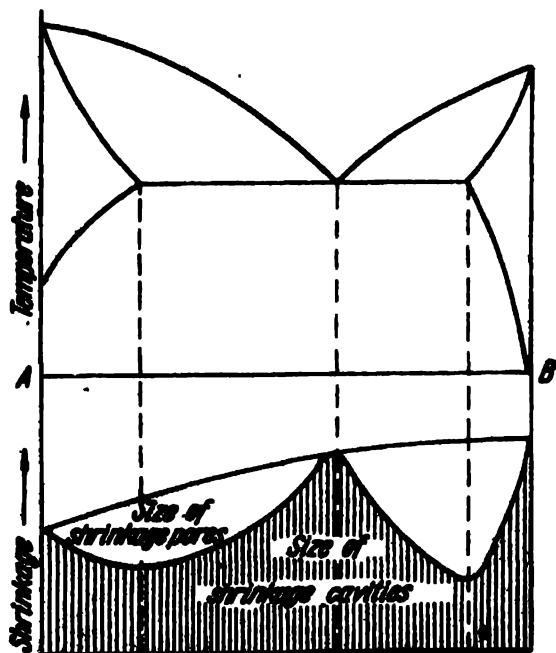

Fig. 37. Change in the volumes of concentrated and dispersed shrinkage cavities as a function of the temperature range of alloy solidification (Bochvar)

The sponginess was measured from the areas obtained with ground unetched longitudinal sections. Although no exact figures

have been found, the measurements yielded an approximate value of 0.5-1.5 mm, i. e., the results have coincided to a degree sufficient for practical purposes.

Proceeding from the theory of alloy contraction developed by A. Bochvar and co-workers, I. Kumanin calculated the shrinkage zone porosity as related to the position of the alloy on the constitutional diagram. The zone of porosity proved to increase at a wider solidification temperature range (Fig. 37). It can be assumed that shrinkage porosity does not occur in castings made of pure metals and eutectic alloys. Here, the entire volume of shrinkage voids manifests itself in the form of a concentrated shrinkage cavity. In actual fact, centreline porosity may develop in thin sections of pure metals and eutectic alloys.
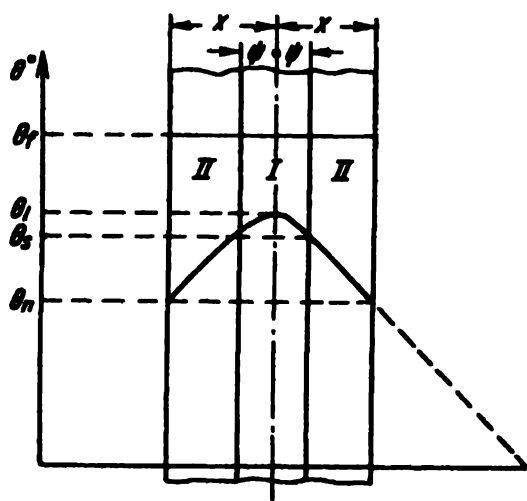
Fig. 38. Change in the thickness 2ψ of the porous zone as a function of the thickness $2x$ in an infinite plate (Kumanin):

I—zone of porous metal; II—zone of solid metal
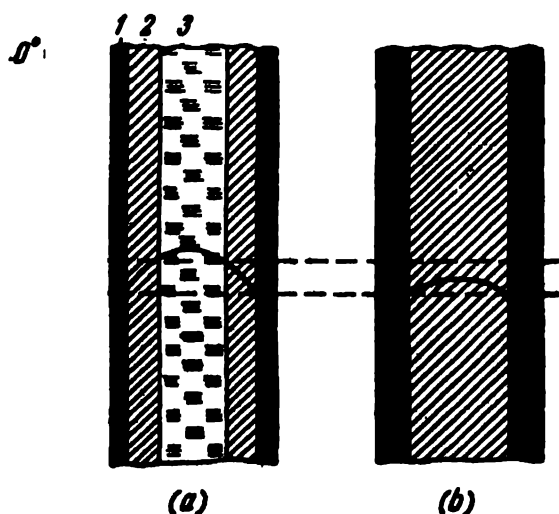
Fig. 39. Stages in the solidification of a plate (Kumanin):

a—first stage—two-phase zone persists (1—solid alloy, 2—two-phase alloy, 3—liquid alloy); b—second stage—two-phase zone disappeared

(a)   (b)

The differential equation of the Fourier heat flow shows that the thickness of the porous zone 2ψ in the casting, in the form of an infinite plate $2x$ thick (Fig. 38), depends on the temperature distribution inside the casting after the mould has been filled.

Solidification should be regarded as a process consisting essentially of two stages. During the first stage of solidification the casting retains an isolated phase of a superheated liquid which can penetrate into the two-phase zone and completely fill the pores which form on solidification. The second stage sets in, when the temperature in the heat centre decreases so much that the zone of the one-phase liquid disappears, as a result of which the reduction in the volume of the two-phase zone can no longer be compensated for by the inflow of liquid from outside. The volume of the solid phase of this zone continues to increase while the liquid phase gradually disappears. Shrinkage pores arise between the crystals. The second stage of solidification is distinguished by the transfer of the superheat from the heat centre and the attainment of the liquidus temperature θ_l by it (Fig. 39). The shaded

Fig. 40. Change in the size of the shrinkage porosity zones depending on the cooling intensity and the temperature range of solidification:

a and *b*—alloys with the same ranges of solidification temperatures but cooled with different intensity; *c* and *d*—alloys with the different ranges of solidification temperatures but cooled with the same intensity

zone, in which solidification occurs when the superheated alloy disappears from the central region, will be porous while the external layers which have completely solidified will be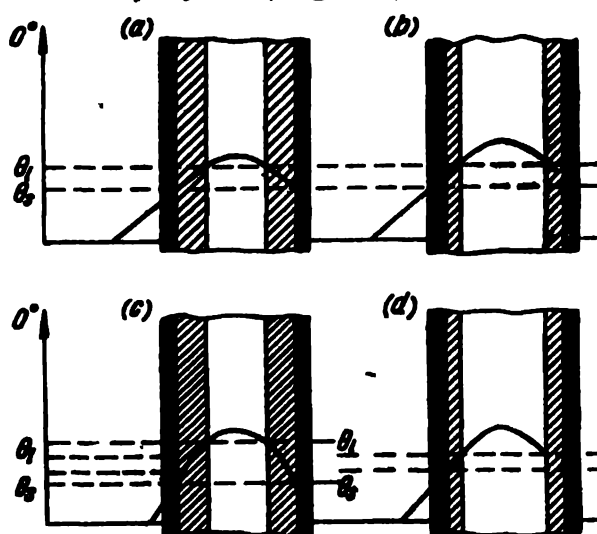 dense. The volume of the two-phase zone, comprised of solid and liquid phases, is determined at the instant θ_l by the temperature range of solidification and the temperature field of the alloy. With a more intensive heat transfer from the metal to the mould, the temperature distribution curve inside the casting will slope at a larger angle and the region of incomplete solidification will be smaller.

The upper part of the diagram (Fig. 40a and b) shows the temperature curves and the shrinkage porosity zones in the castings from the same alloy but cooled with different intensity; in the bottom part of the diagram (Fig. 40c and d) are the temperature curves and the shrinkage porosity zones in castings of different alloys cooled similarly. It follows from this that an increase in the rate of the alloy cooling produces the same effect, in terms of reducing the shrinkage porosity zone, as the decrease in the temperature range of the alloy solidification.

The effect of the solidification temperature range and the intensity of cooling on the development of the shrinkage porosity zone can be found approximately by calculating the size of the two-phase zone by the time τ_1 the centre of an infinite plate, $2x$ thick, is cooled to the liquidus temperature θ_l.

For mathematical analysis, we shall employ Kumanin's equation in its final form

$$2\psi = \frac{2\cos^{-1}\frac{\theta_s}{\theta_l}}{m} \qquad (67)$$

where ψ = half the thickness of the porous zone, cm;

θ_s = solidus temperature of the given alloy, °C;
θ_l = liquidus temperature of the given alloy, °C;
m = solidification constant found from the equation

$$Bi_m = mx \tan (mx), \qquad (68)$$

here Bi_m is the Biot number found from the equation

$$Bi_m = \frac{b_2^2}{\lambda_2 \sqrt{\pi\tau}}. \qquad (69)$$

The Biot number b_2, as reffered to the unit thickness of the plate, can be represented as a function of the heat transfer by the
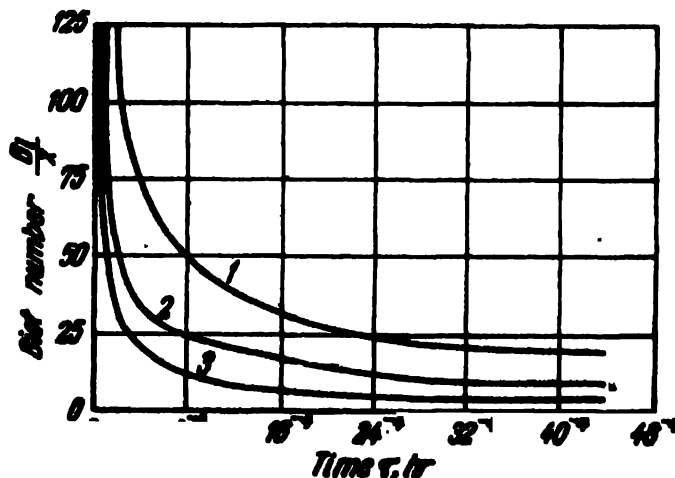
Fig 41. Change in the Biot number and the intensity of heat transfer by the mould as a function of the solidification time:
1—sand mould; 2—green-sand mould; 3—metal mould

mould or, in other words, as a function of its heat accumulating capacity (Fig. 41). For steel castings

$$b_2 = \sqrt{\lambda_2 c_2 \gamma_2}, \qquad (70)$$

where λ_2 = heat conductivity of the mould assumed equal to 20 kcal/m/hr °C;

c_2 = heat capacity of the mould, kcal/kg °C;

$\gamma_2 =$ specific weight of the dense moulding mixture, kg/m³.

For a loam-sand mould $b_2 = 25$ kcal/m³ per hr. For a steel plate 10 mm thick, the Biot number will be 0.12, three seconds after the filling is completed.

The chart in Fig. 42 compiled by I. Kumanin for steel castings illustrates the relationship between the thickness of the porous zone in the casting and its total thickness. On the top is a diagram of the Fe-C alloys structure. According to this chart, only a casting made from pure iron, given a certain intensity of cooling, will possess a solid structure without any porosity. In castings of steel with 0.1-0.4 per cent C the thickness of the porous zone will amount to 0.22-0.55 of the plate thickness. Obviously,
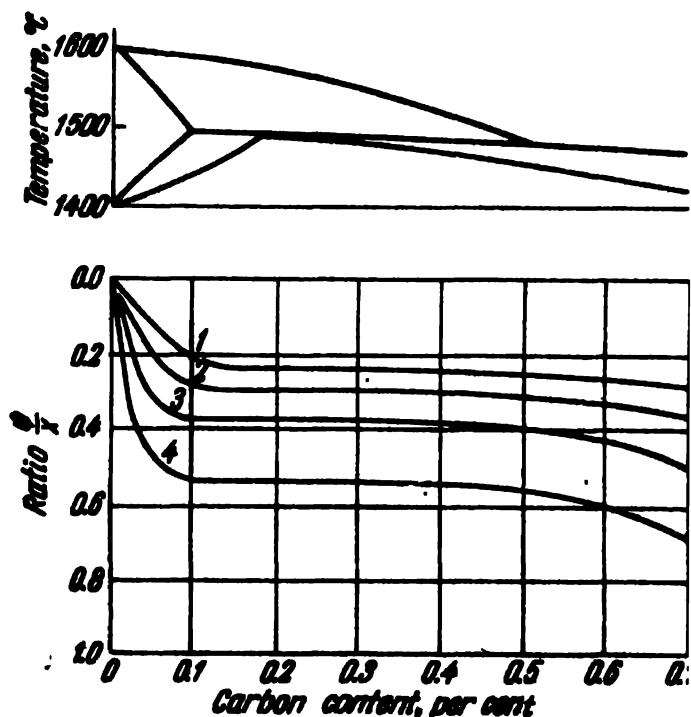
Fig. 42. Change in the Biot number for steel plates depending on the heat transfer by the mould (Kumanin):

1—metal mould; 2—chrome-magnesite mould; 3—green-sand mould; 4—dry-sand mould

in addition to risering, other measures must also be taken to obtain solid steel castings. This will be considered in greater detail below.

13. DIRECTIONAL SOLIDIFICATION

Since any shaped steel casting has an extremely complicated geometry it is rather difficult to determine the relative rate with which the metal solidifies in its various parts. The relative rate

in the growth of the metal solid layer produces a decisive effect on directional solidification which governs the change in the state of the solidifying metal, the position of the one- and two-phase zones, the feeding of an underlayer of metal by the one above and, finally, the obtaining of solid metal throughout the entire volume.

The directional solidification can be most conveniently determined with the help of the A. Heuvers geometrical rule of inscribed circles, illustrated in our case by a junction of a gear wheel rim and a spoke. Let us inscribe a number of consecutive circles into the casting (Fig. 43). The circle 95 mm in diameter indicates to a thickened section in the casting above which circles of smaller diameters (75 and 70 mm) are inscribed.

The riser above the thickened section does not feed it during shrinkage, despite the fact that the diameter of the circle inscribed in it (100 mm) is larger than the diameter of the circle inscribed in the section being fed (95 mm). By providing an allowance as shown in the right part of Fig. 43 and inscribing circles again we shall find that the diameters of the circles are progressively increasing upwards from the section being fed. Calculation of the solidification time, for example by Eqs (16),
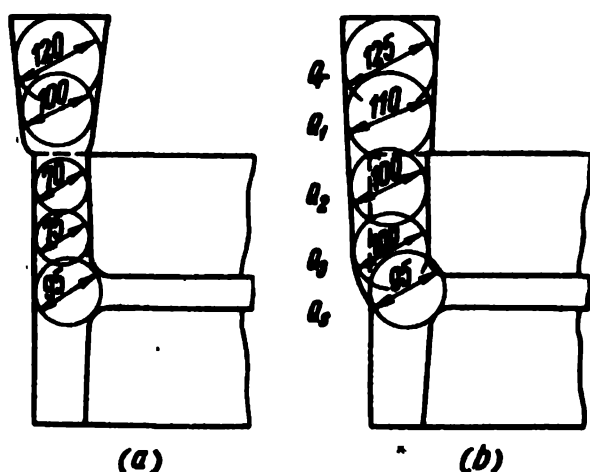
Fig. 43. Application of the rule of inscribed circles in shaping the riser feeding a thickened section in a casting (Heuvers): *a*—wrong; *b*—correct

(a) *(b)*

(46) and (47), which is proportional to the square of the referred thickness R, will show that the section 95 mm in diameter supplied with metal will solidify earlier than the metal on the way from the riser and the riser itself. With the left-side design (Fig. 43a) the liquid metal will solidify near the 70-mm dia circle earlier than in the thickened section and in the riser. As a result, liquid metal will no longer be fed to this section. Conversely, with the design to the right (Fig. 43b) the riser will continually feed metal to this section due to directional solidification.

However, the supply of an underlayer by the one above in the gravity field depends on the difference in the heat content of the adjacent layers. If the heat content of the layer above exceeds that of the layer below the remains of the liquid phase

will move downwards unhampered and close up the pores be-
tween the crystals of the underlayer.

Each successive upper layer will be fed from its counterpart
right above all the way to the riser which should contain a suf-
ficient reserve of liquid steel to be able to feed all the layers
below.

The directional solidification is characterised by the follow-
ing inequality.

$$Q_r > Q_1 > Q_2 > \ldots Q_l > Q_s,$$ (71)

where Q_r = heat content of the riser, cal/kg;
Q_1, $Q_2 \ldots Q_l$ = heat content of the layers below, cal/kg;
Q_s = heat content of the section being fed, cal/kg.

The directional solidification is characterised by the differ-
ence in the heat content of the adjacent layers which should
be within definite limits and be positive for two layers of metal.
If two adjacent layers have the same heat content the directional
solidification will not be secured. When the heat content values
are negative the directional solidification becomes negative, too.
Hence, the second condition can be expressed by the equality

$$\Delta Q = Q_i - Q_{i+1} > 0.$$ (72)

The heat content of any layer of the casting can be calculated
from the formula

$$Q_l = a_i b_i h_i \gamma_i [c_1 (\theta_i - \theta_s) + \varrho] \text{ kcal/kg,}$$ (73)

where a_i, b_i, h_l = dimensions of the metal layer, m;
  γ_i = specific weight of liquid steel, kg/m³;
  c_1 = heat capacity of liquid steel, kcal/kg°C;
  θ_i = temperature of the steel in the i-th layer, °C;
  θ_s = solidus temperature, °C;
  ϱ = heat of solidification, kcal/kg.

Analysing Eq. (73) we can arrive at the dependence of the
directional solidification not only on the difference between
the temperatures of the steel in adjacent sections (Q_i and Q_{i+1})

$$Q \doteq \frac{a}{b}$$ (74)

but also on other factors including the difference in the geo-
metrical volumes of the adjacent layers

$$\Delta V = V_i - V_{i+1} > 0,$$ (75)

where $V_i = a_i b_i$ and $V_{i+1} = d_{i+1} b_{i+1} h_{i+1}$.

Let us consider solidification and freezing of a vertical plate
fed with metal from the top (Fig. 44, A). Since the bottom and
top of the mould walls are raised to the same temperature and
since some hot metal is added to the riser at the last instant the
bottom of the casting will be cooled a certain distance in advance

and the temperature distribution over the casting height will follow the curve a plotted in the height-temperature coordinates along its axis. Soon afterwards the temperature at the lower point 1 lying on the thermal axis will reach the liquidus temperature θ_l. At the next instant of time the curve of the temperature distribution over the casting axis will shift to the right due to cooling and take the position b. The solidification boundary will move upwards and the temperature θ_l will reach point 2 above which the metal will be in a liquid state. At the instant of time marked by the curve c the temperature distribution curve will again shift to the right and the metal of the plate below point $1'$ will solidify completely. From point $1'$ up to point 3 the metal will be solidifying and above point 3 in a liquid state. At the moment d the metal will solidify at the bottom up to point $2'$ (temperature θ_s). Above point $2'$ up to point 4 the metal will be found still solidifying and above point 4 the metal is liquid in the riser. And, finally, at the instant of time described
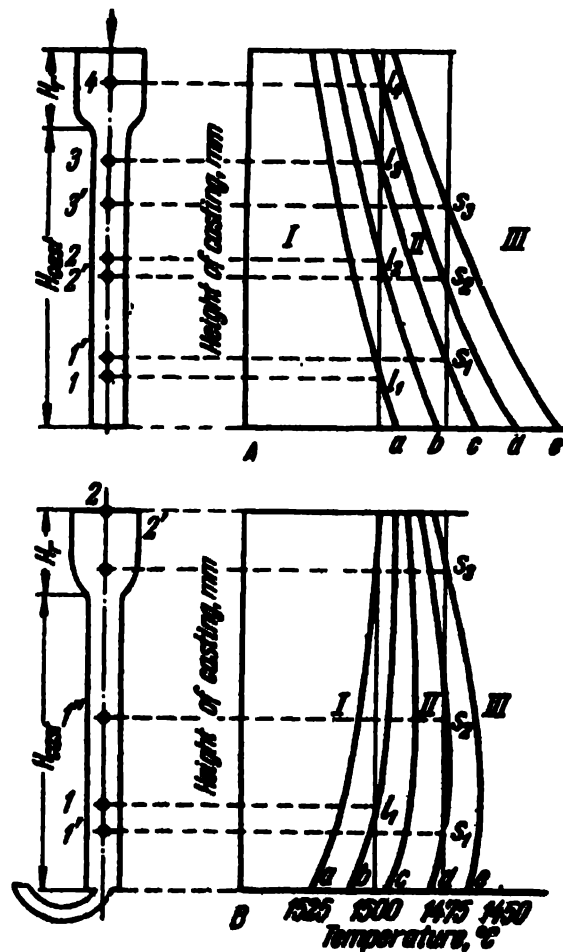
Fig. 44. Longitudinal temperature gradient on cooling a steel plate cast vertically (Bidulya):

A—top pouring; B—bottom pouring; I—zone of liquid alloy; II—two-phase zone over the temperature range of solidification; III—zone of solid alloy

by the curve e the metal will be solidified up to point $3'$ and is still solidifying above it. Thus, when the mould is gated from the top the metal begins to solidify from below with the last portion solidifying in the riser which is the last to solidify.

When the piece is gated at the bottom (Fig. 44, *B*) the hot metal intensively heats the lower part of the mould on its way to the top. At a certain moment of time described by the curve *a* the hottest metal will be below while in the riser it will be the coldest. Allowing for the convection effect (although its possibility may be doubted) we assume that the metal solidifies at a slower rate at the top than at the bottom. Then at the instant of time described by the curve *b* the temperature at the bottom of the casting up to point *1* will be above θ_l and the metal at the top of the casting will begin to solidify. At the moment described by the curve *c* all the metal is in a state of being solidified. There is no directional solidification at that moment. At the moment *d* the temperature distribution curve intersects the vertical line of the solidus temperature θ_s at points s_1 and s_2. Hence, the solidification is upwards above point *1"* lying on the thermal axis of the casting and downwards below point *1'*. At the bottom of the casting the temperature gradient becomes negative. There is no metal fed from the riser and shrinkage porosity will develop in the bottom of the casting. As will be shown below, the riser no longer serves its purpose, not only at the moment *e* but also at the moment *c*, despite the fact that its shape is correct. Apparently, when the temperature gradient is negative, no risering can ensure volumetric feeding of the casting. In other words, the riser will feed the casting during contraction and solidification only when the temperature gradient is positive. The negative directional solidification is distinguished by the fact that during the process of freezing the temperature in some sections of the casting exceeds that of the metal in the riser. The main cause for the appearance of superheated elementary volumes in a casting and for a negative temperature gradient lies in feeding hot metal to the sections far removed from the riser. Hence the following conclusion: to build up a positive temperature difference between the riser and the section being fed the metal should be fed near the riser or poured through the riser. This will ensure the conditions in good agreement with Eq. (74).

The amount of shrinkage porosity in a solidified casting depends on the ratio between the solidification temperatures, the shape of the casting and the position of the mould during pouring.

When the metal is cast with the mould kept horizontal or at an angle with the vertical no centreline shrinkage develops if the risers have been arranged properly and their size is sufficient. Let us illustrate this point by the experiments undertaken by U. S. investigators to obtain solid castings by imparting to them a corresponding sound shape. To get an insight into the factors affecting the formation of centreline shrinkage they experimented

with plates 300 mm high and 6, 12, 25, 50, 75 and 100 mm thick, fed vertically. They sought to determine the effect produced by the relative height of the riser and other factors. The test castings were X-rayed. The composition of the steel in question varied within the following limits: 0.24-0.3 per cent C, 0.52-0.56 per cent Mn, 0.36-0.40 per cent Si, 0.33-0.38 per cent Ni, 0.02-0.03 per cent P and 0.01-0.02 per cent S.

The experiments have yielded a chart (Fig. 45) for the distribution of centreline shrinkage in the top of the plates in per cent of a solid metal as a function of the total height of the casting. Shrinkage was caused by the appearance of bridging which hampered the flow of metal from the riser (not shown in Fig. 45). The maximum shrinkage was observed to develop in thin sections (6, 12 and 25 mm).

If the plates are steadily tapered to the top of the mould or a shrinkage allowance is provided, there will be no shrinkage porosity detected anywhere through the entire volume of the casting. The taper in millimetres per 100 mm of plate height obtained experimentally depends both on the height and thickness of the casting wall (Fig. 46). Thus, for example, for a plate
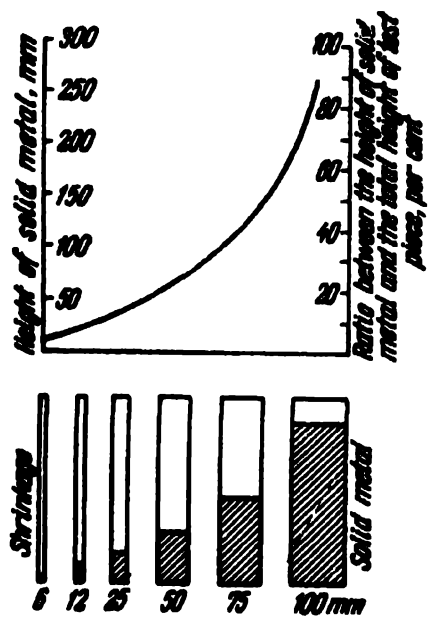
Fig. 45. Effect of section thickness on centreline shrinkage (Brinson, Duma)

25 mm thick and 70 mm high the taper comes to anywhere near 13 mm per 100 mm. At the bottom, the plate thickness equals 25 mm and under the riser it is 13.7 per 91 mm.

Apparently, the shrinkage allowance within the limits shown in Fig. 46 cannot be dispensed with when designing steel castings whose operating conditions rule out completely centreline shrinkage, as for example new cast steel parts for machines, especially from alloy steel which has a broad solidification range and

exhibits the tendency towards the formation of centreline shrinkage.

Although the shrinkage allowance is one of the most effective means to prevent centreline shrinkage, it involves a greater weight of the castings and more time and labour for removing the chips, which should be avoided as much as possible when parts for steam and gas turbines are made from austenitic steel possessing poor machinability. Wedge-shaped plates have been cast, therefore, 400, 800 and 1,200 mm high, 300 mm wide and 70 and 80 mm thick at the bottom. With a different taper the thickness of the plates at the junction of riser and casting was 45-145 mm. This was a dry-sand moulding in a mixture of a chrome-magnesite powder (100 parts by weight), soluble glass with a modulus of 2.3 to 2.5 (7.5 parts by weight), 10-per cent aqueous solution of sodium hydrate (2 parts by weight), and water (2 parts by weight). Grade ЛА-1 steel was melted in a type ДСН-0.5 arc furnace. The metal was gated at the bottom to simulate as much as possible the actual production conditions at iron and steel plants.
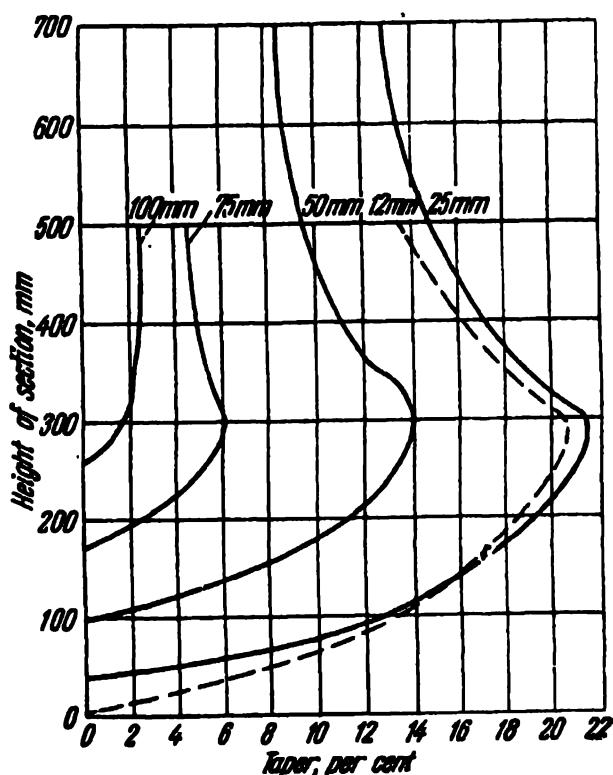
Fig. 46. Taper (shrinkage allowance) in mm per 100 mm of plate height required to obtain solid metal in sections of various height and thickness. Figures near the curves indicate thickness (Brinson, Duma)

The experiments on X-raying wedge-shaped plates and the calculations of the heat content in conformity with Eq. (78) have yielded the interdependence between the taper and specific increment of heat content (Fig. 47).

Further experimental verification has produced this formula for calculating the taper (shrinkage allowance)

$$y = \frac{(\theta_b - \theta_j)(0.1148a - 0.0036)}{0.0574(\theta_b - \theta_j - 2,100) - (\theta_b - \theta_j)}, \qquad (76)$$

where $y=$ taper as determined by the difference in the thicknesses of the wall at the junction of casting and riser and at the bottom of the casting, m;

$\theta_b =$ temperature of steel at the outlet from the bottom gate during pouring, °C;

$\theta_j =$ temperature of steel at the junction of casting and riser during pouring, °C;

$d =$ mean thickness of the casting wall, m.

The lowest temperature θ_b was 1500° C (as indicated by an immersion thermocouple).

When the piece is gated at the top the shrinkage allowance can be reduced twice or thrice as against the design value. With the smallest taper the mould should be filled at a higher temperature and pouring rate applying moulding mixtures with a low heat conductivity.

Practice has yielded yet other methods for improving directional solidification. Here belong faster rates of cooling of the

Fig. 47. Areas of sound steel plates made from austenitic heat-resistant steel as a function of shrinkage allowance and specific increment of heat content (Saveiko)

casting sections far removed from the riser and the heating of the risering elements to slow down their solidification, as compared with the sections of a casting being fed.

These methods, however, only decrease the amount of taper but do not eliminate it altogether.

Different heat transfer conditions can be created through the use of moulding mixtures with a higher heat conductivity, for example, chrome-magnesite or chromite mixtures instead of loam-sand or chamotte mixtures or by pouring the metal into chill and metal moulds.

Risers are usually secured in ordinary nonheat-conductive mixtures while the pouring boxes are made of a dry chamotte

compound. To intensify heat abstraction from the centre of large-size castings, as for example, platens of hydraulic presses, a system of internal chills is used made of iron squares or rounds (Fig. 48). The rough weight of such casting is 143 tons, the weight of the risers—48 tons, of gates—3 tons, and of the whole of liquid metal—194 tons.

Y. Nekhendzi and other authors published data on the dependence of the thickness of the solid layer on the material of the mould (Table 10).

At present, the method of controlling the cooling rate of the mould is being used predominantly for the production of complex castings from high-alloy steels.

It stands to reason that when the rate of cooling of various parts of a foundry mould is controlled artificially, rather solid steel castings can be obtained also with a bottom gate system.
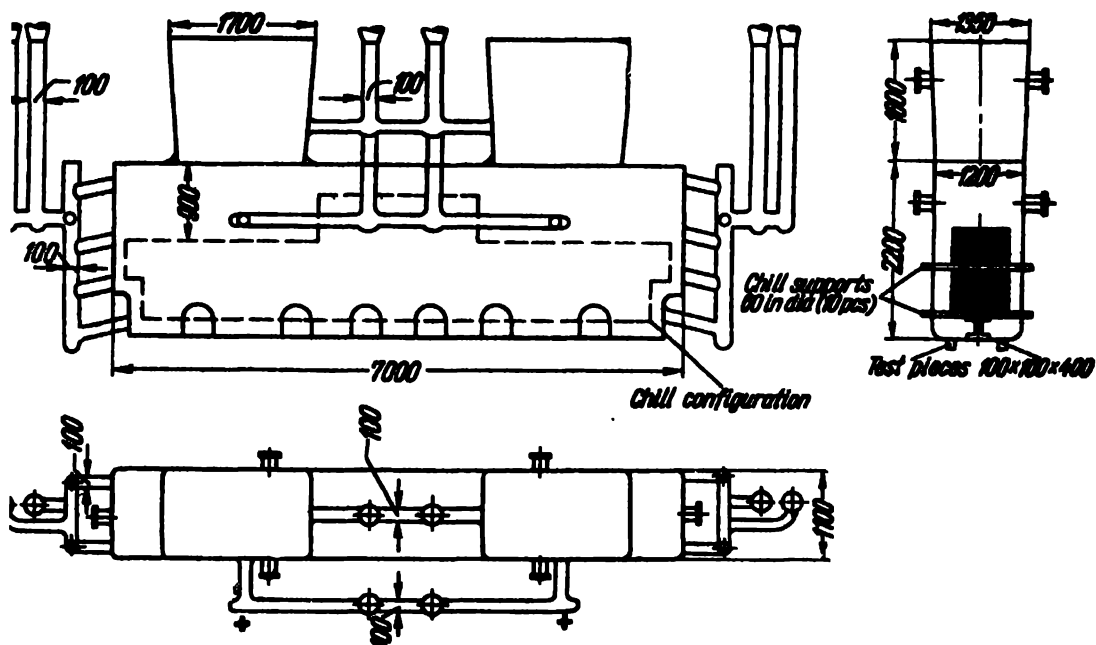

Fig. 48. Feeding a casting for a large-size platen with internal chills (Vasilyev)

The transfer of heat from the riser can be slowed down by the use of heat-insulating facings which will ensure high quality castings and reduce the consumption of liquid metal per riser and consequently increase the yield of sound castings.

The heat-insulating materials for the risers may be porous materials as, for example, light chamotte and foam chamotte, as well as gypsum, asbestos and mixtures of wood flour or sawdust, furnace charcoal ash or ash from the cupola spark arresters. The pouring boxes should be dried and swabbed to eliminate the burning-on on the riser surface. The wash should comprise 6.25 per cent of clay, 57 per cent of silica flour, 1.25 per cent of bentonite, 2.5 per cent of starch gum, and 33 per cent of water.

Table 10

Effect of Various Mould Materials on the Solidification Rate of Carbon Steel

Material	Solidification rate, mm/sec, at a solid layer thickness, mm					
	10	20	30	50	128	180
Copper	10.1	15.2	18.5	26.0	32.0	32.1
Pig iron	8.25	10.5	14.0	21.1	28.6	37.0
Carborundum with grain size, mm						
0.29	7.0	9.25	11.0	16.8	26.0	31.0
0.14	6.6	8.0	10.0	12.5	18.7	25.0
Magnesite	5.25	8.0	9.6	11.0	16.8	21.3
Zirconium	5.0	8.65	10.5	12.0	16.8	23.0
Chromite	4.5	6.5	7.85	10.2	15.0	22.0
Chrome-magnesite	4.9	7.0	8.0	10.4	15.8	20.0
Sand:						
dry	4.0	6.0	7.0	10.0	14.0	19.0
green	3.5	5.6	6.8	9.50	13.5	18.5
Chamotte with a volume weight of, kg/dm³:						
1.2	4.5	7.0	9.5	11.6	16.0	21.0
0.9	3.5	5.0	6.0	8.75	11.0	13.6
0.6	2.5	3.75	4.5	6.0	8.0	10.5
0.3	1.5	2.03	2.5	3.05	4.5	6.5

The mixture for the isolating pouring boxes may be composed of 50 per cent of chamotte, 10 per cent of silica sand, 10 per cent of clay, 30 per cent of saw-dust; the mixture is moistened with soluble glass.

14. HOT CRACKING, CAUSES AND PREVENTION

A hot crack is a narrow irregular crevice the walls of which are covered with dark coloured iron oxide spots developing at high temperatures in the form of hot tears on the surface of a casting. They can also be found inside the castings.

Cracks may also arise at comparatively low temperatures. Their walls have a grainy surface with metallic lustre. These are cold cracks. However, this classification based on a visual appraisal of their walls can by no means explain the causes of crack formation.

Let us consider the mechanism of hot tear formation in a typical case of a restrained contraction of a flat flanged plate

(Fig. 49). The massive part of the sand mould enclosed between the flanges hinders the natural contraction of the plate and the progress of the flanges towards each other in proportion to the mould rigidity. The forces resisting contraction indicated by arrows in the drawing will set up tensile stresses in the plate and shearing stresses at the junction between the plate and the flanges. The total linear shrinkage will change from 1.5 to 0.5 per cent.

According to A. Bochvar, the metal begins to contract the moment a continuous skeleton of the solid phase is formed in the solidifying liquid. But since contraction has already commenced at the moment under consideration after the mould has been filled both the resistance and strain forces will arise simultaneously in the solidified layer.

At the given initial moment the structure of the casting will exhibit a thin solidified skin of continuous dendritic crystals which will progress from the mould walls towards the interior,
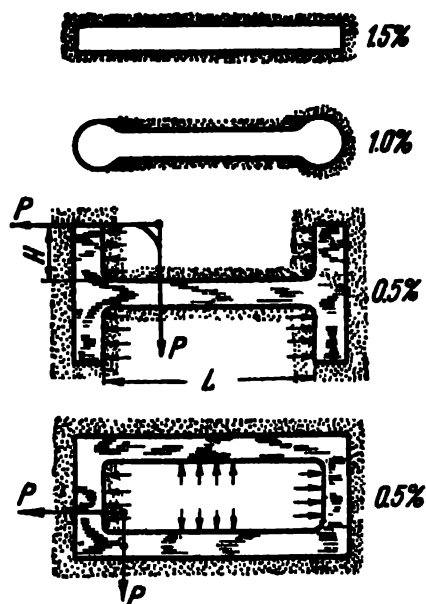

Fig. 49 Formation of a hot tear in a flat steel casting with flanges hindering contraction (Bidulya)

wedging out gradually and passing into liquid. At the start of its formation, the outer skin of the casting is in a solid-liquid state. The layer of liquid is not broken, and if even it does break, it will soon be closed up and the stresses in the solid-liquid layer will begin to increase from zero upwards under the action of the forces resisting contraction. According to Eq. (36), the volume of the skin of the solidified metal will be

$$V = F_1 \xi,$$

whence the thickness of the solid layer is

$$\xi = \frac{V}{F_1}.$$

If the mould is filled almost instantaneously and there is no heating action of the streams of steel fed through the gates,

the solid layer will be uniform in thickness and the stresses caused by restraint of contraction will be uniformly distributed across the section of the entire layer. This is not so in actual practice, since the pouring time differs from zero and the growth of the skin near the gates lags somewhat behind thereby producing different thickness ξ and strength of the solidified layer across the section. The spaces between the dendrite ends will be filled with liquid and will become focal points of stress concentration brought about by restrained contraction.

Fig. 50. Microstructure of a section with a hot crack closed up with liquid metal (Bidulya, Trubitsyn). ×80

For this reason the kinetics of crack formation because of restrained contraction is attended by two processes which occur concurrently: firstly, the mechanical load applied externally, from the rigid mould, increases as the casting is being cooled and contraction develops and, secondly, the thickness of the solid layer and its strength increase during solidification. When the stress exceeds the strength of the skin it will break thereby producing a crack in the casting.

Restraint of contraction may cause a crack to develop in the thinnest section of the skin where solidification lags behind as, for example, near the gates, especially when they are arranged at the heavier sections to improve directional solidification.

The study of casting fractures shows that the metal disintegrates along the grain boundaries because the substance which binds the grains together is apparently weaker than the metal within the temperature range of the solid-liquid state.

If the mould is filled with a superheated metal, the steel may find its way into the crack and close it up (Fig. 50).

The first theoretical research into the causes conducive to the formation of hot cracks in steel castings date back from the

twenties. It yielded only the elementary data of the phenomenon without revealing its physical meaning. Soon afterwards attempts were made for the first time to determine with the aid of a measuring device the magnitudes of critical stresses caused by the restraint of contraction.

Y. Nekhendzi believes that cracks develop in castings at below the solidus temperature, i. e., after the steel has already solidified, mainly between 1450 and 1250°C. B. Osann (Germany) argues that hot cracks may appear even at 600°C. This discrepancy can be attributed, in our opinion, to the simultaneous effect produced by several factors, i. e., purely casting factors (contraction and rigidity of the mould, the size and quality of the skin) and thermal factors (temperature difference in the casting being cooled in the mould).

The temperature difference in the casting depends primarily on the thermal properties of the metal and the material from which the mould is made, i. e., heat content of the elementary areas of the casting and mould towards the end of pouring and the heat-accumulating ability of the mould—the thermal conditions of cooling of a completely solidified casting.

Such cracks may develop, and do actually develop, within a broad temperature range and can be termed annealing cracks to distinguish them from casting cracks caused by the geometry of the foundry mould acting on the metal being cast.

At present, the generally recognised theory of thin liquid films advanced by W. Pellini of the U.S.A. is the most trustworthy theory which explains the formation of casting cracks. According to this theory, a crack is developed by a stress set up due to contraction between the crystals enveloped by a thin film of liquid which separates them. This was confirmed by other investigators in later publications. However, the shortcoming of this theory lies in the fact that the liquid film cannot be broken. K. Singer asserts that the crack will continue to be filled up as long as a flow of liquid capable of increasing the solidifying layer is provided from inside. The crack, therefore, takes its final shape somewhat later, when the inflow of liquid suddenly ceases due to the loss of fluidity, i. e., when the liquid steel near the crack attains the temperature of zero fluidity.

The failure is caused not by the presence of liquid films around the crystals. If such films exist, the whole of the crystalline aggregate cannot be possible in a solid state and there is nothing to be disintegrated. The drawing of the adjacent crystals apart by contraction will cause a fresh inflow of liquid into the intercrystalline spaces. Obviously, the crystals will finally detach from each other when the intercrystalline liquid phase will solidify completely near the cracks.

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The resistance to hot cracking near the solidus temperature depends apparently on the strength of the intercrystalline layer. The admixtures active on the surface which build up monomolecular adsorption layers on the surface of the crystals as well as new phases can essentially interfere with the intercrystalline bonds. The nature of adsorption depends on the elements present in the steel. For example, an addition of aluminium or chromium to nondeoxidised steel will reduce the surface activity of oxygen and increase the resistance to cracking. The treatment of liquid steel with small additions, but within a rather narrow limit of their relative weight, is one of the most effective methods to govern the adsorption processes and, hence, resistance to cracking. Such additions include cerium, boron, calcium and other deoxidisers.

Numerous attempts have been made of late by various investigators to find the temperature and the stresses conducive to the appearance of cracks due to restrained shrinkage.

The moment of failure can be registered with a device which makes it possible to establish the critical temperature of the thermal centre, or hot spot, close to the temperature of zero fluidity but lying between the latter and the solidus temperature. This device, designed by the author in cooperation with N. Trubitsyn, V. Gruzin, I. Avilov and V. Saveiko (Fig. 51), permits a simultaneous recording of the stress which tears apart the entire test piece and its subsequent drop registered with the help of a recording electronic potentiometer of the ЭПП-09 type with six or twelve loops.

The graduated odd loops of the device are intended to take temperature readings and the even ones—to measure the stress acting on the test piece and caused by restraint of contraction. One end of the test piece (Fig. 52) is rigidly secured with an invar bolt while the other is connected to a flexible segment like the spring in the device developed by Briggs and Gezelius. The data units are quite original in design. They transform linear movement during contraction into an electrical pulse by means of a wire strain gauge consisting of a strip of paper with a thin looped constantan wire glued on it. When elastically deformed by the steel segment the wires of the strain gauge become longer and change its electric resistance counted by a Wheatstone bridge. One diagram can be used to record the readings of several tests poured simultaneously or at definite time intervals. The thermocouple is enclosed under a thin quartz cap and placed in the centre of the test piece where the tearing has become troublesome.

This device has been used to establish the relationship between the resistance to cracking (critical stress producing a crack) in steel of various grades and the pouring temperature (as indicated by an immersion thermocouple) (Fig. 53). As the pouring
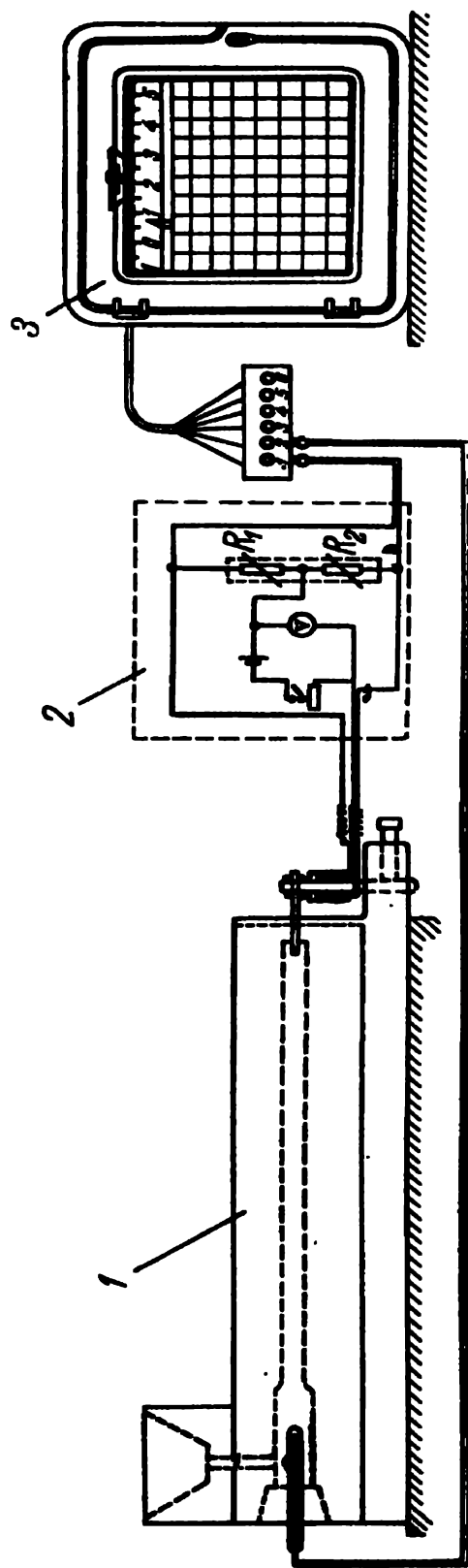
Fig 51 Device for measuring the stresses and temperatures during the formation of a hot crack in restrained contraction (Trubitsyn, Bidulya et al):

1—moulding box; *2*—intermediate switch; *3*—type ЭПП-09 potentiometer

temperature is raised, the resistance to cracking sharply increases and reaches its maximum for the steel of each grade over a definite temperature range, after which it drops sharply and exhibits the tendency to growth only at a high superheat. The low temperature of the steel during pouring (almost without superheat) is unable to close up the cracks due to contraction at its very start. Apparently, the ability of a superheated steel to resist crack formation increases and reaches its maximum at a 2-5 per cent superheat. Higher superheat noticeably reduces
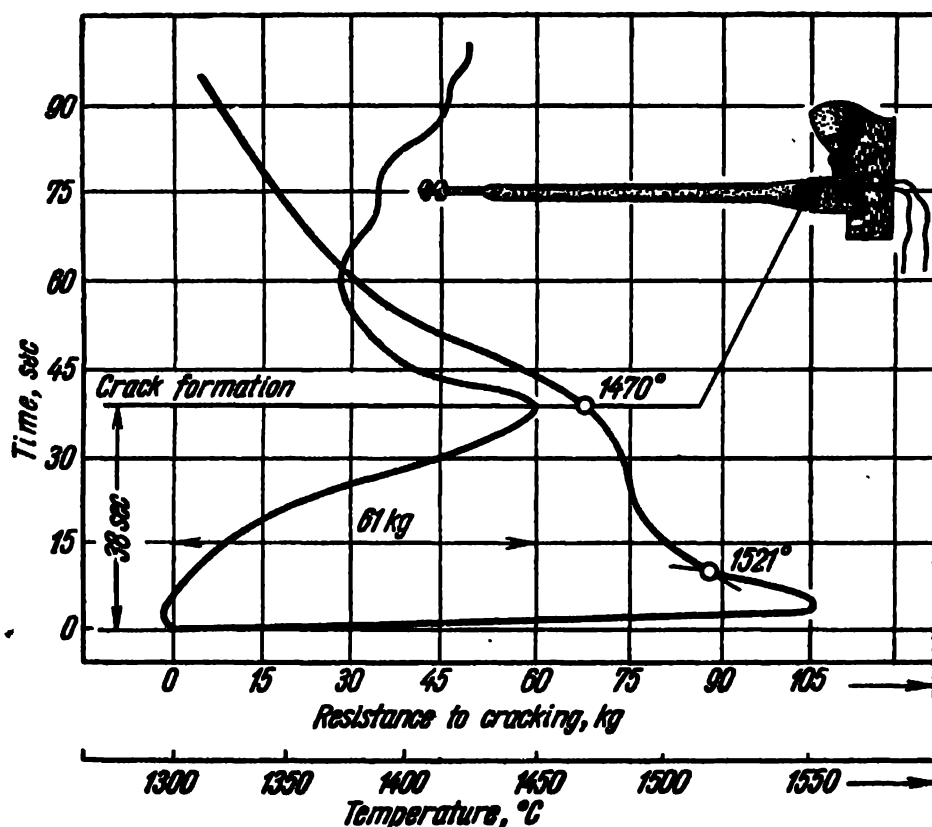

Fig. 52. Test piece for determining resistance to hot cracking of steel and the curves of the temperature (right-hand curve) and stress (left-hand curve) during solidification attended by restrained contraction (Trubitsyn, Bidulya)

the resistance to cracking. This can be accounted for by a considerable temperature difference between the hot spot and the periphery and by the formation of a thinner skin on the casting which takes over the entire stress from the restraint of contraction. An increase in the test piece diameter in excess of 30 mm shifts the curves of the maximum resistance to cracking to the left, towards the lower pouring temperatures, and a decrease in the wall thickness to 15 mm shifts them to the right.

The effect produced by the pouring rate and the temperature on the resistance to cracking of grade Л40 steel was especially pronounced in the experiments in which the metal was fed from

a different location (Fig. 54). The metal was fed near and aside from the hot spot. A gate arranged at the heavier section (Fig. 52) supplies a large amount of heat and heats the test piece so thoroughly that the solidification of liquid metal in the heavier section is retarded. Even when the mould is filled at a slow rate the resistance to cracking comes to about 20 kg, whereas when liquid metal is fed at the thin section and at a slow rate the resistance of steel to cracking increases to 200 kg.

Fig. 53. Resistance to cracking versus stress (in kg) at which the test piece is torn apart (Trubitsyn, Bidulya):

I—grade 25ХМЛ steel; *2*—grade 25Л steel; *3*—grade 25ХМФЛ steel
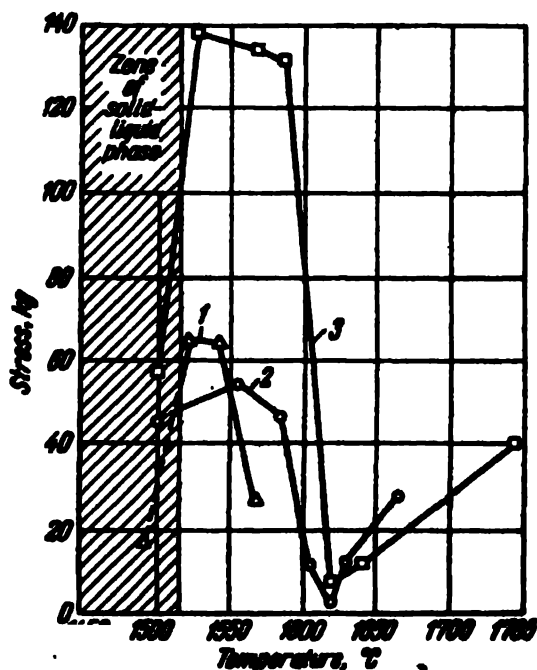
These experiments helped to gain an insight into the kinetics of hot crack formation under restrained conditions. Considering a test piece with a heavy section as a stressed structure, we will notice that if the hot spot is heated with the metal supplied directly from the gate inserted into the spot, this spot will be the most vulnerable, because the formation of a solid skin will be somewhat retarded. After a solid crystal skeleton has been formed on the surface of a thin section, the amount of linear shrinkage will gradually increase. At this instant, the skin in the heavy section is so thin that it breaks when contraction stresses build up in it, then is closed up again, is torn anew, etc.

Finally, this process ceases altogether when the surface of the crystals is coated with oxides of sufficient thickness which prevent further closing up and the hot tear will stop in its development if the tensile stress is removed all of a sudden. If the stress is not removed the crack will pass right through the casting material.

When the metal is fed to a thin section it will give off part of the superheat to the mould walls, as it fills the mould, and for this reason will be somewhat colder at the heavy section.

The solidification of the skin at the heavy section will be intensified and freezing will proceed over a larger area. The weakened section and stress concentration will disappear. The curves of linear shrinkage and uniform stress distribution will take another course (white triangles in Fig. 54).

Sometimes the test pieces will develop no hot tears, i. e., the metal will remain solid in all sections and over the entire length.

There is so far no generally accepted idea of the effect of steel composition on resistance to cracking. It has been found experimentally that the maximum resistance distinguishes steel with 0.20 per cent C (Fig. 55). Further increase in the carbon content first reduces the resistance to cracking and then slightly improves it.
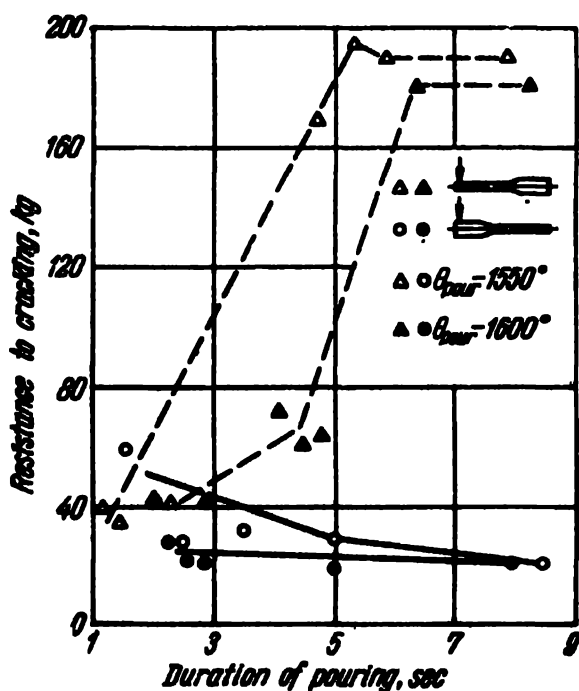
Fig. 54. Change in the resistance to cracking (dash curves) and stresses (solid curves) depending on the location of the source feeding metal to the test piece and duration of pouring (Trubitsyn, Saveiko):

white symbols—metal fed to heavy section, black symbols—metal fed to thin section

N. Girshovich, Y. Nekhendzi and B. Lebedev investigated the resistance to cracking of Fe-C alloys within a broader range of carbon content (0.1-3:9 per cent) and confirmed our conclusions. On the other hand, K. Beckius (Sweden) has arrived at different conclusions.

In his opinion, steel with 0.2 per cent C possesses the minimum resistance to cracking and steel with 0.4 per cent C—the maximum resistance.

Resistance to cracking can be accounted for only by the nature of the bond between the crystals in the hard skin which will have solidified by the time the stresses have developed due to restraint of contraction. It can be easily seen if we subject a ribbed test piece to a tensile test (Fig. 56). The thin ribs which solidify at a fast rate are distinguished, firstly, by a fine grain

structure as distinct from the structure of the hot spot containing larger crystals and, secondly, by the fact that the thin section of the ribs will have cooled and strengthened the vulnerable place of the casting being tensioned by the time the increasing stress due to restrained shrinkage has fully developed. This phenomenon discovered long ago has been utilised as one of the principal measures to prevent hot cracking in steel castings.
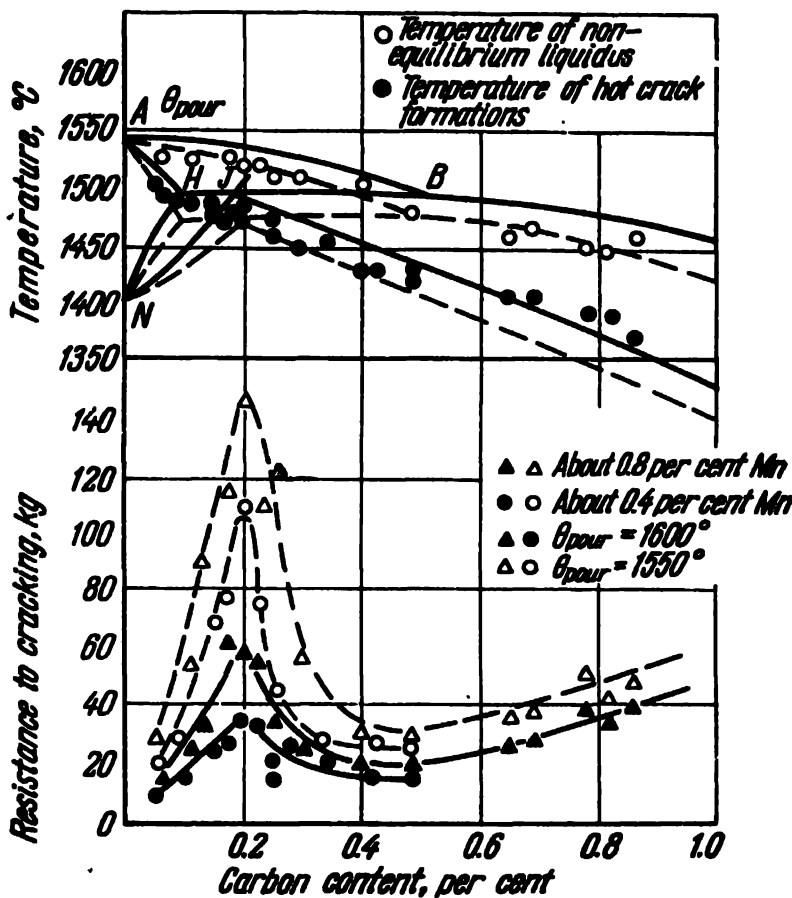
Fig. 55. Change in the steel resistance to cracking as a function of carbon content and pouring temperature (Trubitsyn, Bidulya)

I. Lupyrev and B. Gulyaev studied the effect of the hot spot and the chills on the change in steel strength after the mould has been filled. The diagram in Fig. 57 indicates the steel strength

Fig. 56. X-shaped test piece for determining steel resistance to cracking (Bidulya)
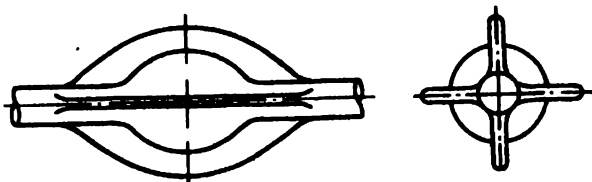
at a rupturing temperature of test pieces depending on the cross section and the rate of stress increase brought about by restrained contraction. For example, forty seconds after filling,

the minimum strength was observed in piece *2* with a cross-shaped section and the maximum strength in piece *5* with a T-shaped section from which the solidified reinforcing rib is cut off after the casting is knocked out.

The heating of the hot spot produces a reverse effect. For example, several blades of a hydraulic turbine were discarded as rejects at the Elektrostal Plant (U.S.S.R.) because hot cracking had developed under the flange on which a massive riser has been previously installed. The reinforcing ribs cut under the flanges during moulding were of no avail. Good results were
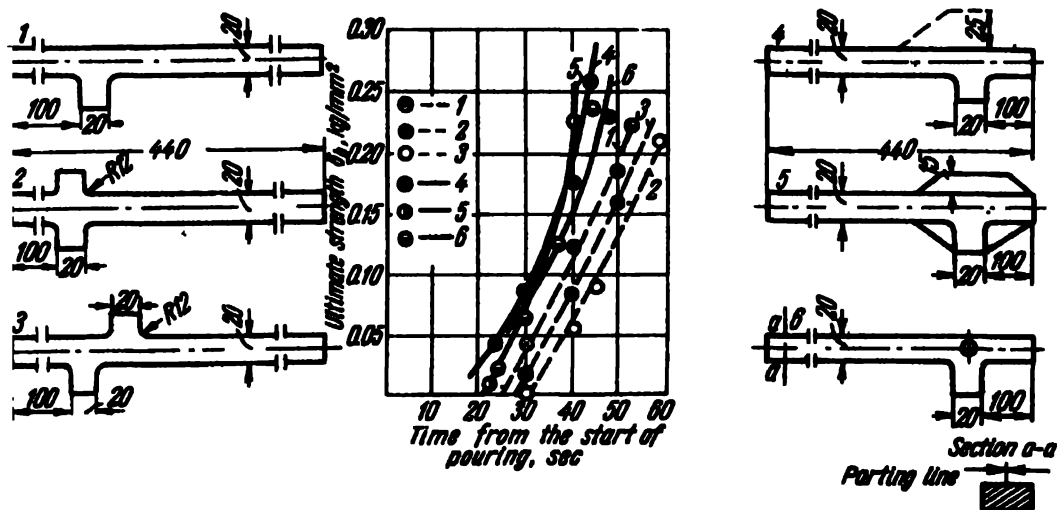
Fig. 57. Effect of the hot spot and the chills on steel strength after pouring (Lupyrev, Gulyaev)

obtained with a new method which provided for higher rates of the solid skin growth by placing a water-cooled pipe into the mould.

Admixtures in steel which increase the temperature range of solidification, such as sulphur forming fusible sulphides, are regarded as tending to decrease the resistance to cracking. An increase in the sulphur content above 0.03 per cent may in many unfavourable circumstances sharply reduce the resistance to cracking. A larger amount of manganese neutralises the action of sulphur due to the formation of refractory sulphides and increases the resistance to cracking. This can be seen on the diagram in Fig. 58 compiled from the data of numerous experiments.

Alloying the steels with much manganese (for example, Hadfield steel, grade Г13Л) reduces the resistance to cracking because of a broader temperature range of solidification. Even an insignificant warpage brought about by the solidification of thinner gates can produce hot cracking in this steel, as was demonstrated by A. Ryzhikov. Such cracks were prevented by

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Fig. 58. Effect of admixtures in carbon steel on the resistance to cracking at the pouring temperature with the same superheat (50°C above liquidus) as a function of chemical composition (Trubitsyn, Bidulya)
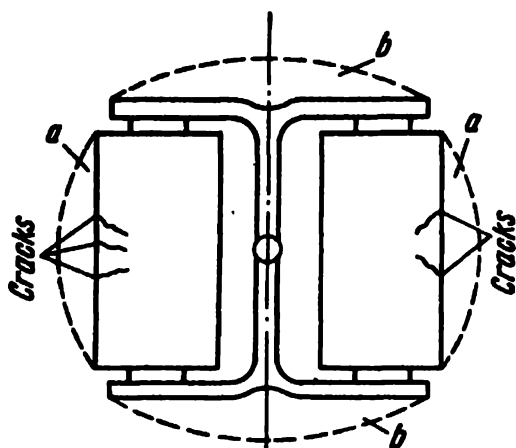
Fig. 59. Casting of grade Г13Л steel with reinforcing ribs a and b (Ryzhikov)
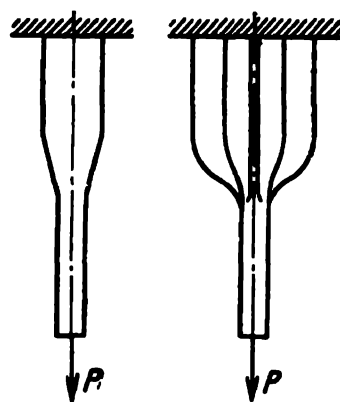

Fig. 60. Reinforcing ribs to prevent cracking in a steel casting (Saveiko)

Table 11

Dimensions of Reinforcing Ribs and the Distance Between Them (Saveiko)

Thickness of the casting main wall, mm	Rib thickness, mm	Height of rib h_p, mm, at the ratio of the arm length of the casting horizontal wall to the thickness of the vertical wall				Distance between the ribs, mm, at the ratio of the thickness of the vertical wall to the thickness of the casting horizontal wall		
		up to 8	over			up to 1	over	
			8-16	16-24	24		1-2	2-3
6-8	2.5	10	9-10	12-16 *	16-20	35	30	20
8-10	3.5	7	10-13	15-20	20-25	50	40	25
10-12	4	8	12-15	18-22	22-30	60	50	30
12-16	5		17-20	23-30	30-40	80	65	45
16-20	6	12	20-24	28-36	36-48	100	80	55
20-25	7	15	25-30	35-45	45-60	110	90	60
25-30	8.5	17	28-34	40-50	50-70	130	100	70
30-40	11	20	35-45	50-60	56-85	170	140	90
40-50	12	25	45-50	60-75	75-100	190	150	110
50-60	14	30	50-60	65-85	85-115	200	160	110
60-80	16	35	55-70	75-100	100-135	220	180	120
80-100	18	35	60-75	85-110	110-145	240	190	140
100-140	21	40	70-80	90-125	115-170	250	200	150
140-180	23	45	75-90	105-135	135-180	250	200	150
180-220	25	50	85-105	120-155	155-205	250	200	150
220-300	30	60·	100-120	140-180	180-240	250	· 200	150

means of cracking strips in the form of thin reinforcing ribs (Fig. 59).

The section of reinforcing ribs, their arrangement and number can be determined by calculation. Suppose the lower part of the casting (Fig. 60) with a heavier section is subjected to tension due to restrained contraction. The upper end of the casting is tightly secured and a weight P is suspended from the bottom end. The heavier section carries n reinforcing ribs. We assume that solidification lags behind in the heavy section and the force P is taken over only by the thin section of the casting and the ribs. We can therefore write the following equation:

$$n\sigma_r F_r = \alpha\sigma_m F_m, \qquad\qquad (77)$$

where n = number of reinforcing ribs;

 σ_r = specific stress across the rib section, kg/cm^2;

  F_r = section across the rib, sq cm;

  α = coefficient accounting for the increase in the strength in the massive part of the castings accruing from the reinforcing ribs;

  σ_m = specific tensile stress across the thin section of the casting, kg/cm^2;

  F_m = section across the thin section of the casting, sq cm.

Fig. 61 Standard designs of reinforcing ribs (Saveiko):
a—angular; b—longitudinal; c— curved or filleted
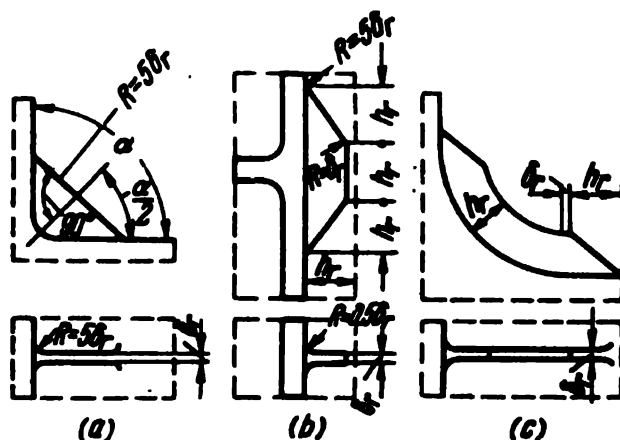
As a rule, the thickness of reinforcing ribs varies between 0.1 and 0.3 of the thickness of the casting thin section. With the maximum restrained contraction $\alpha=0.25$ and with the minimum $\alpha=1.00$. The rib height-to-thickness ratio is 8 in the first case and 2 in the second. Standard designs of reinforcing ribs are represented in Fig. 61 and their dimensions and the distances between them in Table 11.

In conclusion we shall describe certain measures to prevent hot cracking in the order of their effectiveness.

1. The causes of restrained contraction should be eliminated, for example, by providing cavities in the mould near projections.

Such cavities form after removing wooden blocks during moulding (Fig. 62). During shrinkage the projections on the casting shift and break off the moulding mixture causing unrestrained contraction.

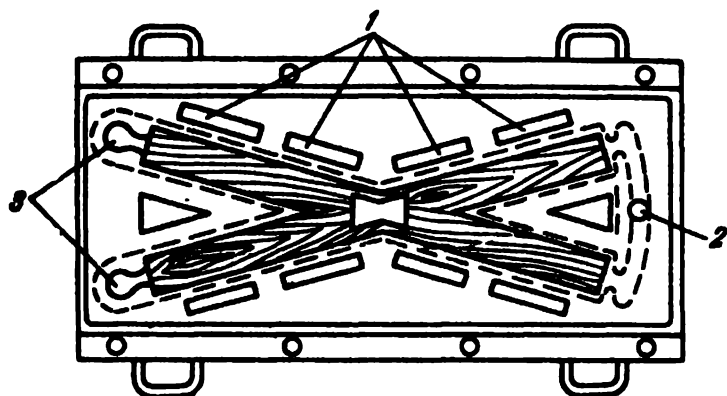Fig. 62. Mould with cavities left during moulding to collapse the mixture and preclude restrained contraction:

I—cavities: *2*—sprue; *3*—overflows (Goncharov)

2. An attempt should be made to design all sections in a casting with uniform thickness to avoid abrupt changes in section.
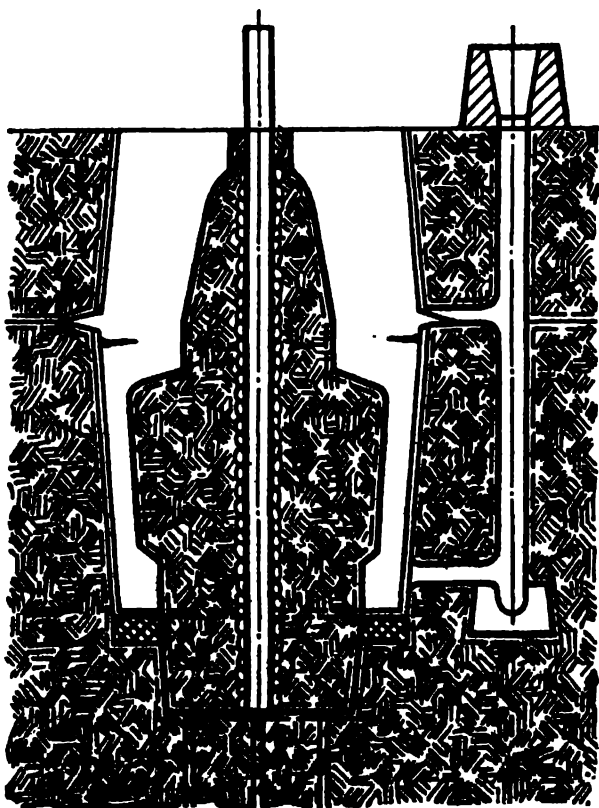

Fig. 63. Hanging of a vertical casting on fins during contraction and the formation of a crosswise hot crack (Bidulya)

3. It is desirable to ensure good directional solidification to preclude heat concentration in the intermediate hot spots and to feed metal away from these heat centres, striving to chill rather than heat them.

4. The metal should be poured into the moulds at a moderate rate and at a reduced temperature.

5. Excessive sulphur content, if unavoidable, should be neutralised with manganese.

6. Provision should be made for local chills and reinforcing ribs.

7. Measures must be taken to prevent hanging of the casting during contraction (Fig. 63).

15. RESIDUAL STRESSES, WARPING AND ANNEALING CRACKS

Contraction and temperature gradients during solidification and freezing give rise to thermal stresses which entail various undesirable phenomena. If the value of local stress exceeds the yield point of metal the dimensions of the casting may change, i. e., it will be warped. If these stresses grow above the ultimate strength the casting may develop cracks. With the stresses below the yield point three kinds of forces may arise in the casting which will be equalised:

(a) within the volume of the entire casting;

(b) within one crystal;

(c) within one or several crystal lattices.

The stresses of the first kind are vitally important for steel foundry practices.

The thermal stresses are brought about by nonuniform cooling and different rates of solidification and contraction in various portions of complex and even the simplest casting or ingot. The phase transformations which occur over the temperature range of pearlitic transformations as well as restrained contraction, which does not allow high stresses to appear and cracking to develop, may be the cause of residual stresses.

The magnitude and sign of internal stresses can be controlled in production conditions. When the casting process takes a correct course such stresses can be reduced to the extent that their consequences will not affect the quality of ready-made products. However, to be able to control residual stresses in castings made of carbon steel we must know how to determine their magnitude with registering devices, for example with a magnetic anisotropy indicator. This device operates on the following principle: if an alternating current enters a massive metal body with a flat surface via a straight conductor in a direction perpendicular to this surface then, if the body is iso-

tropic, the magnetic lines of force will propagate in concentric circles lying in a plane parallel to the surface of the product.

Magnetic anisotropy distorts the magnetic field in the product and near the point where the conductor is joined. This generates dispersion fluxes intersecting the body surface. The intensity of the magnetic field grows in proportion to the amount of anisotropy. Since an alternating electromagnetic field arises only in the surface layers the dispersion fluxes can be used to determine the anisotropy of these layers.
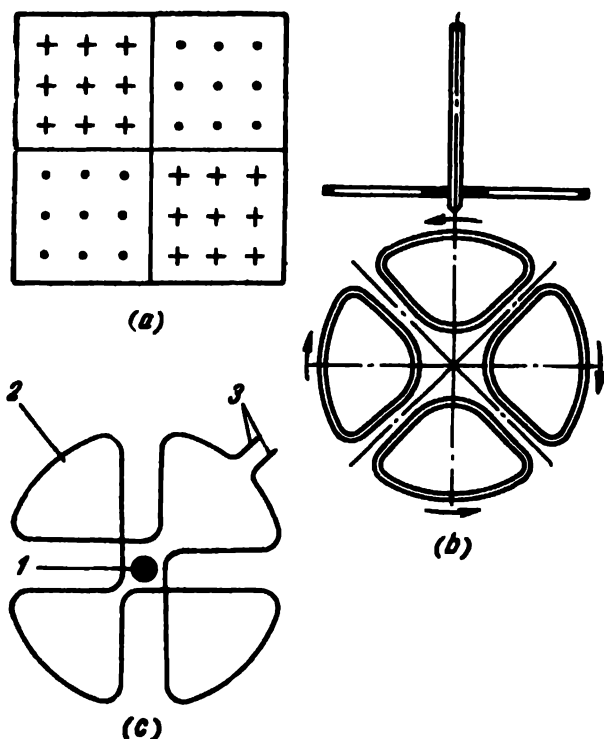
Fig. 64. Device for measuring magnetic anisotropy (diagram):
a—direction of dispersion fluxes; b—location of the coils and electrode; c—arrangement of the turns

Let us assume that the current flows from the observer towards the plane of the drawing (Fig. 64). In this case the direction of the dispersion fluxes can be represented as shown in the diagram a. The directions of magnetic anisotropy plotted on the drawing through the connecting point of the conductor located in its centre will divide the whole field into four squares forming intermediate neutral lines along which the magnetic induction of the dispersion fluxes is zero. One of these lines coincides with the direction of the elastic surface stress and the maximum value of magnetic permeability.

By recording the intensity and direction of the magnetic dispersion fluxes as well as the location of the neutral lines we can establish not only the degree of magnetic anisotropy but also the orientation of its maximum values. For this purpose use is made of type ИМА-1 magnetic anisotropy indicator designed by B. Timofeyev.

The device comprises four flat coils (Fig. 64b) wound with numerous turns of thin copper wire 0.02 mm in diameter. The

coils are tightly pressed against the surface being investigated and a current electrode is placed between them. If the system is so arranged that one of the lines of the main direction of magnetic permeability passes between any two adjacent coils the electromotive force between the ends of these coils will be at the maximum.

By pressing the device against a flat ground and polished surface of a casting and passing current through the electrode and the casting we can measure the residual stress set up in the casting. The device is graduated by means of a strip $(8 \times 34 \times \times 1,000$ mm in size) made from the steel being tested. One end of the strip is fixed and a weight is suspended from the other

Fig. 65. Graduation stress curve of type ИМА device (Shiryaev, Bidulya, Novikov)
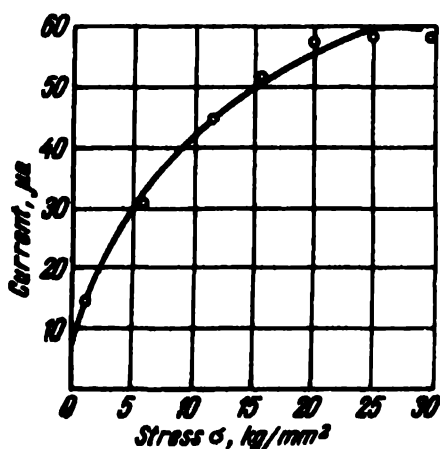
which twists the strip and stretches its fibres. By measuring the stress at the various points of this strip we can plot a curve of the type shown in Fig. 65. This curve is used to evaluate the device readings when it is applied to the actual casting of the same metal. The accuracy of the emf values determined with the aid of this device amounts to ± 3 per cent. The values of residual stresses in a plate $500 \times 1,000 \times 5,000$ mm in size, as measured with this device at three points on the surface, are illustrated in Fig. 66.

Let us now consider how thermal stresses arise in a casting which is cooled in a foundry mould down from a high temperature, using a simplified scheme. The casting can be regarded as a bundle of bars made integral. For the sake of simplicity we assume that the casting consists so far of only two bars: thick *1* and thin *2* (Fig. 67). During the pouring process their temperature will be the same. But on cooling, after the time z_1 elapsed from the end of pouring, the temperature of thin bar *2*, following along the lower temperature-time curve, will be below that of thick bar *1*. The temperature change diagram can be used to plot the dependence of the contraction and, hence, of the linear dimensions of the bars on the cooling time (Fig. 68). Let us refer the change in dimensions to unit length $(l=1)$, at a temperature
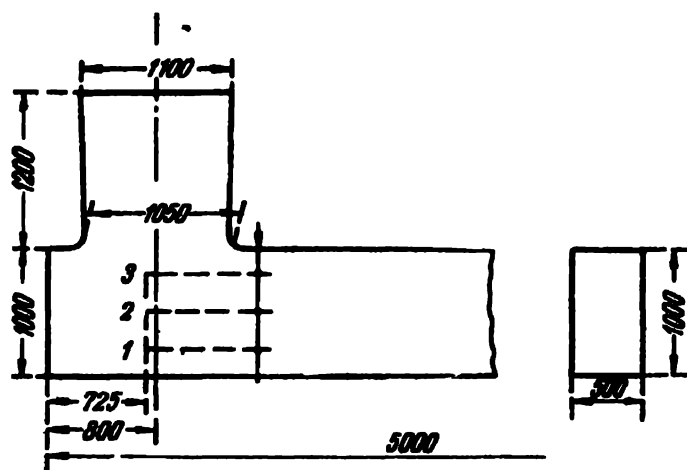
Fig. 66. Measuring residual stresses in a heavy steel plate caused by cooling down from a high temperature (Shiryaev, Bidulya, Novikov):

1-3 — points of measurement

Point	Before annealing		After annealing	
	mean, mv	kg/σ, mm^2	mean, mv	kg/σ, mm^2
1	36.0	7.5	16.4	1.8
2	40.6	9.5	24.0	3.5
3	41.6	9.8	25.0	4.0
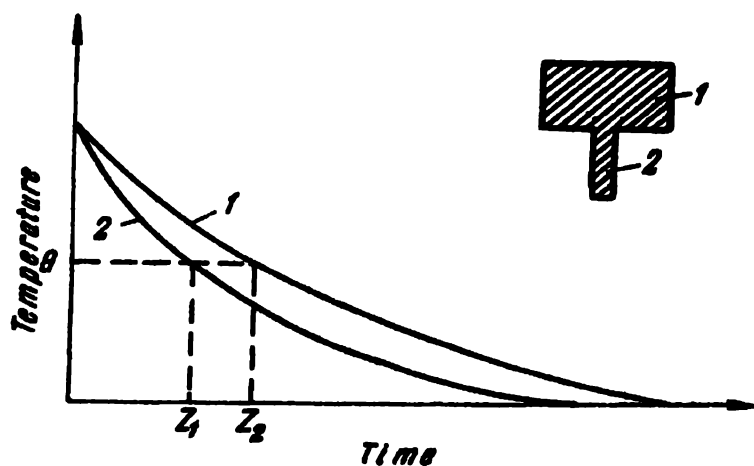
Fig. 67. Change of the temperature in thick *1* and thin *2* parts of the bar depending on the duration of cooling in a foundry mould

$\theta=0$. The coefficient of linear shrinkage being denoted by α, the length of any bar can be expressed by the equation

$$l = l_0(1-\alpha\theta), \qquad (78)$$

where l_0=initial length of the bar, cm;
$\quad$ α=coefficient of linear shrinkage;
$\quad$ θ=temperature difference between the initial and present state of the system, °C.

After selecting a corresponding ordinate scale for our diagram (Fig. 68) we replace temperature by a proportional length.

After the time z_2, the length of thin bar 2, which cools at a faster rate than bar 1, will be equal to b_1d_1 and the length of bar 1 to a_1d_1 if both the bars are not united in a common system. But since they constitute a single whole—the casting—its mean length will be equal to the section c_1d_1. The length of the thin bar will increase by the amount c_1b_1 and that of the thick one will decrease by a_1c_1.

If the whole of the casting is in a plastic state the thin bar will develop a plastic tensile deformation and the thick one— a plastic compressive deformation and there will be no stresses set up in the casting.

Let us assume that the bars are in a plastic state during the time τ_1. Then the length of the thin bar will increase by the amount c_1b_1 and that of the thick bar will diminish by a_1c_2. The

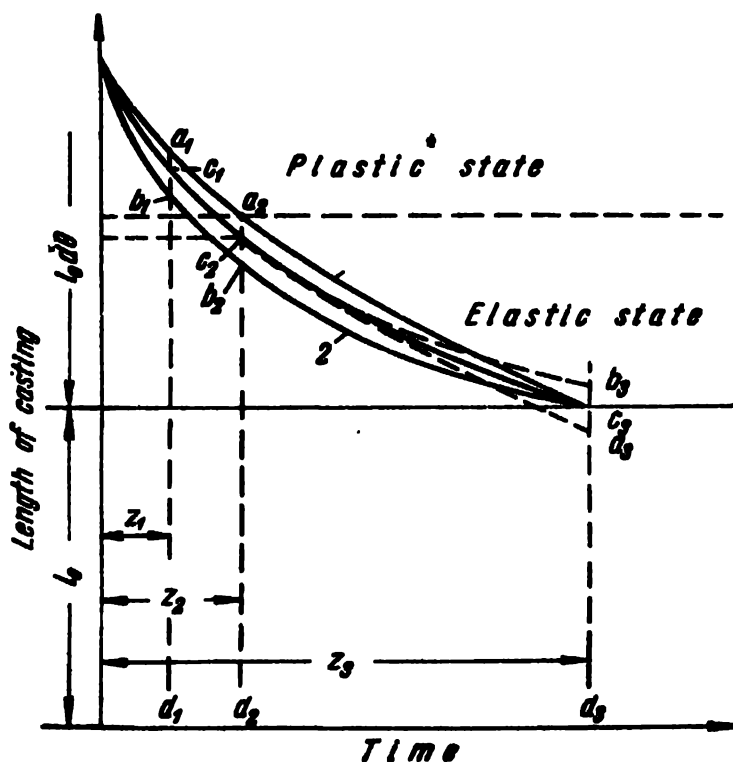*Fig. 68.* Change in the length of the thick and thin parts of a casting depending on the duration of cooling in a foundry mould

casting will undergo plastic deformation which will eliminate the stresses.

Let us further assume that after the time τ_2 thin bar 2 has become elastic while the thick one still retains its plastic state. Then, reasoning as before, the plastic deformation will decrease the length of bar 1 by b_2c_2, while the length of bar 2 will become equal to the mean length of the entire casting due to elastic deformation, i. e., to c_2d_2. If at this instant the two bars are separated the length of bar 1 on natural contraction will correspond to the section c_2a_3 and that of bar 2 to the section b_2d_3. After the time τ_3, when the bar has cooled to room temperature, the length of thick bar 1 on independent natural contraction will become equal to the section a_3d_3 and that of bar 2 to the section b_3d_3. Both bars are in an elastic state. But in actual fact they are united into a single rigid system whose mean length will equal the section c_3d_3. Obviously, bar 1 will be in a state of elastic tension by the amount of c_3a_3 and bar 2 in a state of elastic compression by the amount b_3c_3. Instead of l_0, the length of bar 1 will be $l_0 - c_3d_3$ and that of bar 2 will be $l_0 + c_3b_3$.

The total elastic deformation in the system comprised of bars 1 and 2 can be found from the equation

$$\varepsilon = a_3c_3 + b_3c_3 = a_2c_2 + b_2c_2, \qquad (79)$$

i. e., it will equal the amount of plastic deformation arising when the heavy part passes into the region of elastic deformation but will be inverse in sign to plastic deformations. The tensile stresses are denoted by the plus sign and the compressive stresses by the minus sign.

If bars 1 and 2 retain their straight shape the stresses arising in them can be expressed by the ratio

$$\frac{\sigma_1}{\sigma_2} = \frac{f_2}{f_1}, \qquad (80)$$

where σ_1 and σ_2 = stresses in bars 1 and 2, kg/mm²;
f_1 and f_2 = sections across bars 1 and 2, sq mm.

Let us assume that the moduli of elasticity on compression and tension are the same, then

$$\sigma_1 = E\varepsilon_1 \quad \text{and} \quad \sigma_2 = E\varepsilon_2, \qquad (81)$$

$$\frac{E\varepsilon_1}{E\varepsilon_2} = \frac{f_2}{f_1},$$

$$\frac{\varepsilon_1}{\varepsilon_1 + \varepsilon_2} = \frac{f_2}{f_1 + f_2}. \qquad (82)$$

The sum of the deformations $\varepsilon_1 + \varepsilon_2 = a_3 c_3$, i. e., it is equal to the total deformation at the moment when bar 1, and consequently the entire system, passes into an elastic state; the total deformation in this case will be equal to the difference of deformations of both bars from the moment bar 1 passes into an elastic state until the entire system cools completely.

Let us denote the temperature of bar 1 at the moment of change-over into an elastic state by θ_1 and that of bar 2 by θ_2. Then their difference will be

$$\theta_1 - \theta_2 = \Delta\theta, \tag{83}$$

where $\Delta\theta$=the temperature drop between bars 1 and 2 at the moment the heavy bar has passed into an elastic state.

The sum of stresses will then be

$$\varepsilon_1 + \varepsilon_2 = \alpha\Delta\theta, \tag{84}$$

where α=coefficient of linear expansion or shrinkage.
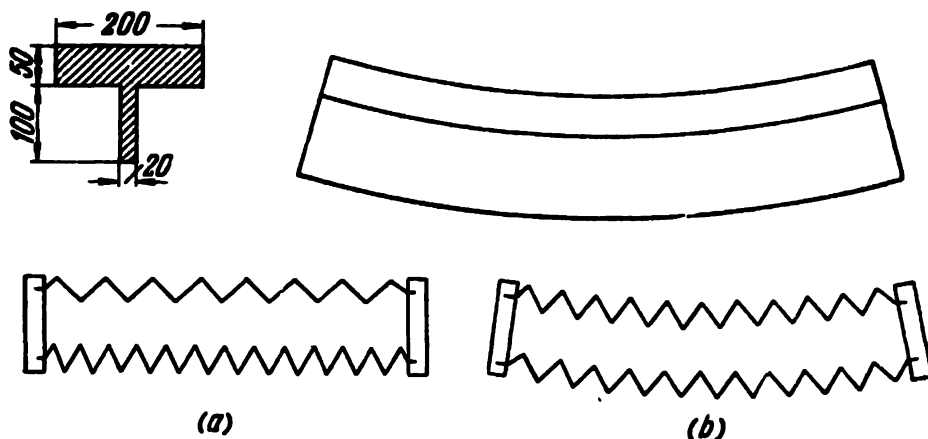
Fig. 69. Deformation of a T-shaped casting on cooling:
a—before cooling; b—after cooling

Substituting in Eq. (82) the right-hand side of Eq. (84) instead of $\varepsilon_1 + \varepsilon_2$ we obtain

$$\varepsilon_1 = \frac{f_2}{f_1 + f_2}\alpha\Delta\theta. \tag{85}$$

The stresses in bars 1 and 2 can be determined from the equations

$$\sigma_1 = E\frac{f_2}{f_1 + f_2}\alpha\Delta\theta \tag{86}$$

and

$$\sigma_2 = -E\frac{f_1}{f_1 + f_2}\alpha\Delta\theta. \tag{87}$$

It follows from Eqs (86) and (87) that the tensile and compressive stresses in the castings are proportional to the modulus of elasticity, the temperature difference between the thick and

thin sections of the casting, at the moment the thick section passes into an elastic state, and the coefficient of linear expansion. The stresses increase in proportion to the difference in the cross sections of the casting, which determines the temperature difference. In thick sections cooling at a slower rate the stresses become tensile and in the thin sections with a faster rates of cooling they become compressive.

The tensile and compressive stresses deform the entire system. For example, a T-shaped bar (Fig. 69) represents, as it were, two springs secured between two blocks (Fig. 69a). If the entire system is relieved of stresses (Fig. 69b) the bar will be deformed.
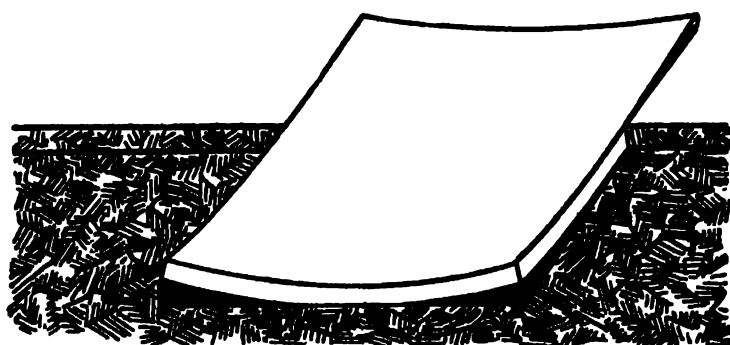
Fig. 70. Warping of a flat plate cast in an open floor mould (Bidulya)

By measuring the temperature of a similar steel casting being cooled, we can determine approximately the stresses which arise in the thick and thin sections. When the thick section of the casting is raised to 650°C and is entirely in an elastic state, the temperature difference comes to about 90°. The modulus of elasticity for steel with 0.2 per cent C is 21,000, and the coefficient of linear expansion can be assumed to equal $1.8-10^{-5}$. Then, according to Eq. (86), the stress in the thin section of the casting will be

$$\sigma_1 = 21,000 \ \frac{20 \times 100}{50 \times 200 + 20 \times 100} \ 1.8 \times 10^{-5} \times 90 = 5.67 \ \text{kg/mm}^2.$$

From Eq. (87) the stress in the thick section of the casting is

$$\sigma_2 = -\sigma_1 \frac{f_1}{f_2} = -5.67 \ \frac{10,000}{2,000} = -28.3 \ \text{kg/mm}^2.$$

Warping and cracking must be prevented first of all by a proper design of a casting. Abrupt changes between thick and thin sections or sharp corners should be replaced by fillets. There should be no concentration of metal in large-size castings whose warpage cannot be corrected.

It is a frequent case in practice when soundly conceived castings have to be rejected in the process of designing. What with the need to correct the shapes of such colossal structures as crossbeams for hydraulic forging presses weighing hundreds of tons, it will become clear why a thorough and critical approach to the design is quite indispensable.

Nonuniform cooling of various parts of large plates (Fig. 70) may be a frequent cause of warping. When the plate freezing in a mould is uniform in thickness its centre is cooled last (see Fig. 69). The stresses caused by tensile forces are not arranged symmetrically since the upper surface of the plate cools faster

Fig. 71. General view of an annealing crack on a steel casting of a cylinder cap from the formation of thin fins on the edges causing a high temperature difference

than the lower. This will result in a casting with a swelling turned downwards. Sometimes annealing cracks develop in the centres of such unsound structures as large-size flat castings (Fig. 71).

The loosening of the mixture in the centre of a foundry mould with a casting at a yellow heat temperature (1000-1100°C) tends to mollify somewhat annealing cracks and reduce warping. Thick plates are usually cast in a vertical position.

Steel possesses both plastic and elastic properties.

For example, a steel ball dropped from a certain height onto a white hot steel plate rebounds from it as from an elastic body, despite the fact that steel can be forged and rolled as a plastic material at such temperatures. At room temperatures steel also yields to rolling or drawing. Purely elastic properties distinguish hardened steel containing over 0.3-0.4 per cent C. Thus, in a general case, steel within the range of 1540-0°C and below

can be classed among the materials with partially elastic properties. Consequently, all the above considerations concerning the causes and conditions under which annealing cracks form in steel castings can without much error be attributed not only to the formation of cold but also of hot cracks, according to the old classification.

This can by no means be applied to hot casting cracks caused by restrained contraction and mould geometry.

Cracks that form when castings or ingots hang on fins are ranked among the casting flaws. Cracks frequently observed at the junction between a heavy riser and a thin section of the casting below the riser develop at high temperatures and cannot be classed among casting cracks since they are caused mainly by the temperature difference which is of a thermal rather than of casting nature.

The following conclusions can be drawn:

1. Thermal stresses in castings cooling down from a high pouring temperature are caused mainly by the different rates of cooling in different volumes of the casting. From the standpoint of heat engineering, the stresses are the result of temperature difference between various parts of the volume.

2. A linear dependence exists between the stress and temperature difference. The smaller is the temperature difference near red heat, the smaller are the stresses and, consequently, warping and cold cracking are less probable.

3. The stresses decrease in proportion to the difference between the sections of the thick and thin parts of a casting.

4. The faster the centre of a thick casting freezes, the less probability for the formation of tensile stresses and cold cracking.

The following measures are instrumental in preventing warping and cold cracking in steel castings.

1. Smooth transitions from thick to thin sections.

2. Faster rates of cooling of heavy sections, for example, with the aid of external chills and of the centres of heavy castings with internal chills. This by no means dispenses with the methods of ensuring proper directional solidification to remove centreline porosity caused by contraction. On the other hand, an excessively pronounced directional solidification is rather harmful. Much harm accrues from thin fins in the form of thin projecting ridges of metal forced out where the mould halves do not fit together properly. This will inevitably lead to cold cracking in heavier sections of a casting due to a considerable temperature difference. On the other hand, reinforcing ribs provided at the heavier sections, especially at the juncture between the thick and thin sections, will prevent the development of hot and cold cracking.

3. Greater dimensions of a casting require a more uniform heating of the mould walls, by feeding the metal through many gates for example. Concentration of heat near one gate will tend to increase the temperature difference between the section near the gate and the adjacent thin sections and may be the cause of cold cracking.

The analysis of the causes liable to produce temperature difference in a casting being cooled rules out completely such unsound large-size complex castings in which the hazard of a considerable temperature difference cannot be avoided.

CHAPTER III

MANUFACTURE OF STEEL CASTINGS

All the technical operations involved in making steel castings, as well as castings from other ferrous and nonferrous alloys, may be categorised as belonging to one of the following fundamental steps of the process.

(a) Preparation of drawings and process charts. The working drawings are supplied by the designing office, the drawings of moulding riggings and process charts are the responsibility of the engineering department while the time standards are provided by the rate setting department. The time schedule for the manufacture of separate castings or batches of castings is set by the production and the planning departments for a month or shorter periods of time.

(b) Patternmaking (including core boxes, templates, moulding boxes and other auxiliary equipment). Patterns and core boxes made in the pattern shop are then shipped to the foundry where their cast counterparts are produced. They are further processed in the machine or the tool department.

(c) Preparation of moulding and core sands.

(d) Moulding. Moulding consists of all operations necessary to prepare a mould to receive molten metal.

(e) Melting and pouring.

(f) Cooling and shaking out of castings.

(g) Cleaning to remove moulding and core sands.

(h) Cutting off the gating, risering and fins.

(i) Preliminary inspection of the castings for defects and general quality. Defective castings may be salvaged by welding or other repair.

(j) Heat treatment.

(k) Final inspection.

(l) Finishing operations.

(m) Shipment.

Foundrymen must bend every effort towards obtaining high quality castings at the lowest cost and within the shortest terms possible by means of a properly organised and planned production process.

The engineering department should be amply provided with up-to-date scientific and reference literature, process charts

and standards usually compiled by each plant locally. Much depends on the production experience and thorough knowledge of the latest advances achieved by foundrymen at home and abroad.

The technical process adopted for casting a die bedplate at the Kramatorsk plant (U.S.S.R.) may serve a good illustration of how the accuracy of large-size steel and iron castings can be increased and the moulding rates speeded up.

For the casting use was made of chemically solidifying moulding and core sands mixed with soluble glass. The solidification process was carried out by blowing the packed sand with carbon dioxide before the mould or its parts were separated from the pattern.
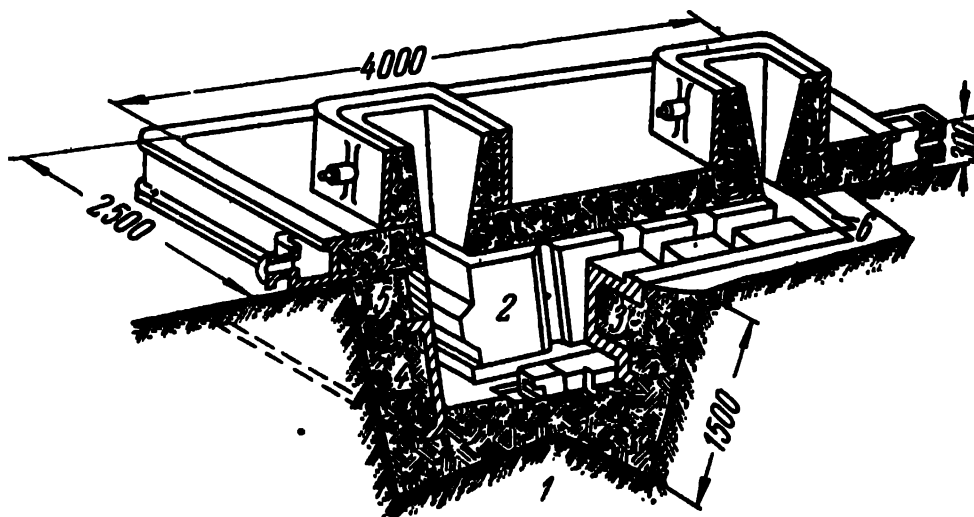
Fig. 72. Mould for a die bedplate with shell cores and inserts (Kestel):

1—core; 2-6—shells

The mould was assembled from five disjointed shells (Fig. 72) containing 80 per cent of silica sand and 20 per cent of mar-shalite, adding to the mixture 7.5 per cent of soluble glass of specific gravity 1.48-1.52 and a modulus of 2.2-2.4 and 1 per cent of sodium hydrate (10-per cent solution). The compressive strength of the green sand mixture was 0.25-0.35 kg/cm^2; the dry tensile strength was 7 kg/cm^2 and the dry compressive strength 20 kg/cm^2. Permeability amounted to 80 with the moisture content of 3.5-4.5 per cent. The shells were 50-60 mm thick. The carbon dioxide was blown for 8-25 min at 2.5-3 atm.

The bottom of the bedplate was cast by a frame-shaped pattern. Core 1 was inserted into the mould print seat, with a template secured on its top. Shells 2 and 3 were set close to the template and the distance between them thoroughly checked. Then, shell 4 was inserted into the seat and checked whether it was perpendicular to the bedplate axis. Shell 5 was placed into the

endfaces of shells *2* and *3*, parallel to shell *4*, and checked with the template. The shells were secured with studs.

Used sand was placed and rammed outside the shells and chamotte pipes laid to serve as gating channels. Shell *6* was inserted into the upper row of the gates. A moulding box was fixed and made tight along the upper planes of shells *2* and *3*, with runner bushes secured on its top. The mould was weighted down with 60 tons and then filled with grade 35Л steel. Some coal dust was put into the riser which was then filled with molten metal to the top. The allowances accounted for only 5-7 mm instead of 12-18 mm, which reduced the amount of chips, sub, sequently removed, from 756 to 314 kg, i. e., by 58.5 per cent. Much less time was needed to prepare the mould than before.

16. PROCESS DESIGNING

Process designing is the main concern of the engineering department in a foundry or the chief metallurgist office at a plant and ranks among the most important processes in production. In his work, the process designer relies on the working drawing of the part to be cast, the production schedule of the shop and the technical conditions for the manufacture and shipment of cast products. Process instructions, State standards and reference handbooks are used as ancillary aids.

The working drawing indicates the name of the casting, its purpose, all the necessary dimensions, the grade of accuracy of the working surfaces, the finish weight and the grade of steel as well as all the other recommendations of the designers. The working drawing should be regarded as a production order of the designer to the process engineer and has, therefore, legal force. However, the working drawing becomes a legal document only after the specific features of the design have been discussed by the designer and the process engineer responsible for its manufacture. The drawing should bear the signatures of both the designer and the process engineer, which are regarded, in a way, as a pledge of the design soundness.

The design of a hub developed by Czechoslovak foundrymen shown in Fig. 73 may serve a good illustration of a sound construction. This design is distinguished by a high quality, simplicity and convenience in manufacture, low cost and highly reliable operation. Unfortunately, in actual practice, not all of these requirements can be fulfilled completely and the moulds usually fail to function perfectly in one or more respects because designers sometimes lack practical production experience.

Most unwelcome designs appear sometimes, involving heavy and ineffective outlays. Shaped steel castings for powerful machines should be preliminarily checked for soundness at the

experimental shop or plant. The study of such castings (or numerous test batches) will fully pay for themselves since valuable corrections can be made in the design during testing which simplify the technical processes and enhance the reliability of the design. Essential corrections must be approved, of course, by the chief designer.

When designing the pattern and foundry mould conventional signs are noted down on the working drawing to show how to make the pattern and core boxes and assemble the mould and how to feed liquid metal to the mould to obtain a uniform casting.

Fig. 73. Position of a steel hub in pouring (Pišek):
a—wrong; *b*—correct
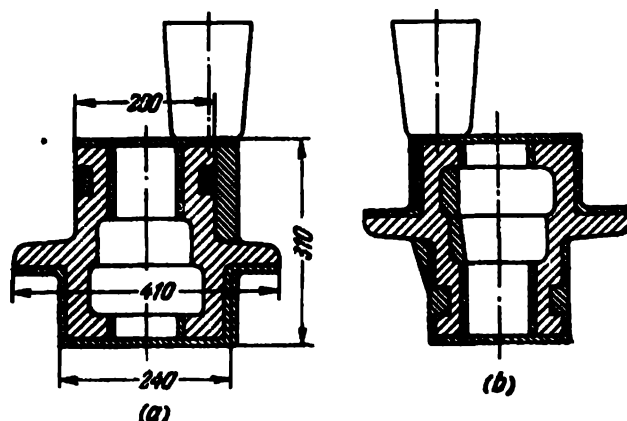
The drawing is supplemented with a process chart which specifies in detail the course of the foundry process for the knowledge of foremen, team-leaders and inspectors from the technical inspection department.

The process designing of a foundry mould and riggings should be preceded by the selection of the mode of manufacture which depends on the possibilities of the shop and the kind of production of the castings. The designers must decide whether the castings will be machine- or hand-moulded, and what method to use—green-sand or dry-sand moulding.

When the moulds are filled from a stopper ladle the pressure head of the molten metal is sufficiently high as is the probability of sand wash. For this reason, provision should be made for mould drying and the use of moulding sand mixed with soluble glass and cement. In conveyor-type foundries green-sand moulds are ordinarily used. If such a mould with a hygroscopic core is allowed to remain idle for more than 5-7 hours before use the castings may develop gas holes. This may also cause the chills and studs in such green-sand mould to grow damp.

As an additional operation, mould drying requires more fuel, handling facilities and time. When the mixtures are prepared with soluble glass the drying of moulds and cores may be frequently dispensed with.
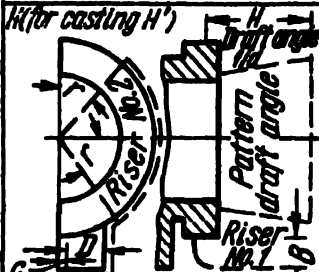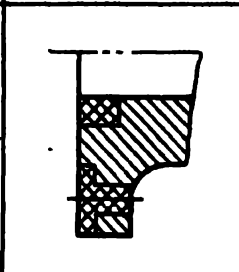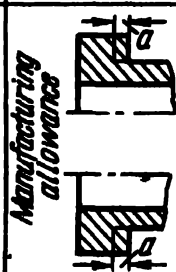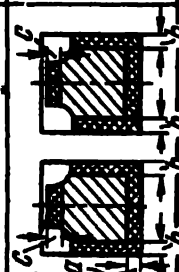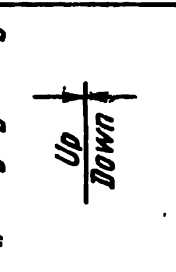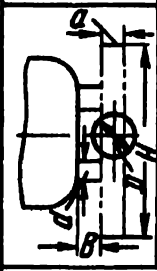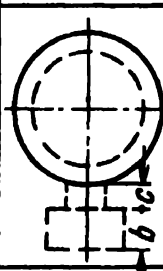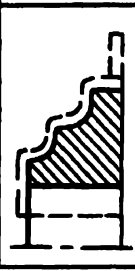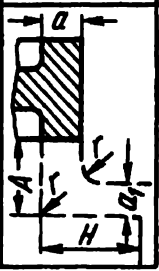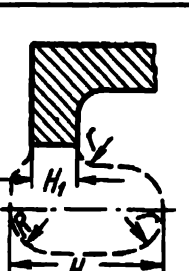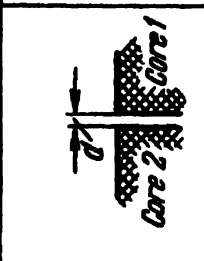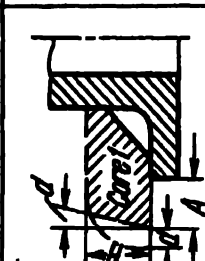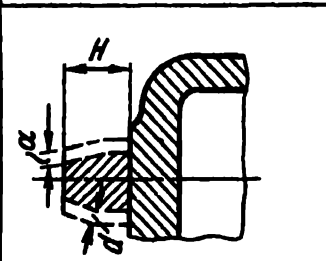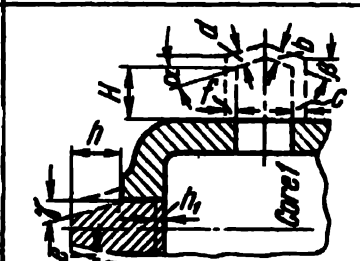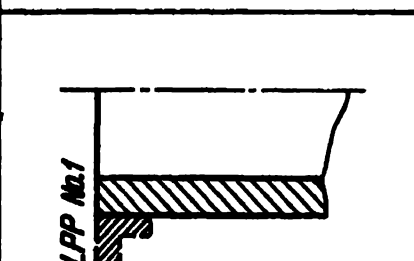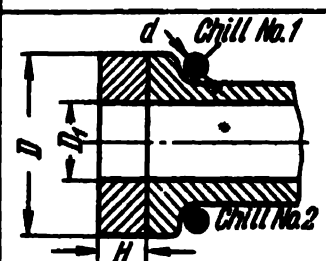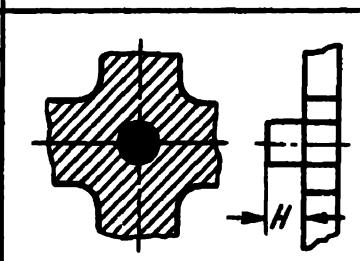
Riser	Holes, slots and shoulders not to be cast	Manufacturing allowance	Machining allowance	Parting plane

Gating system	Shrinkage ribs	Location and size of sprue	Test piece	Template	Flow-off	Branch-off riser

Core boundary	Side of ramming in	Core print outline	
		Not in section	In section

Loose piece of pattern	Chills	
	external	internal

Fig. 74. Conventional signs on part drawings used in process designing

Hand moulding of large castings is an extremely time-consuming process and is performed in dry-sand moulds.

The next step is to select the position of the casting in the mould so as to ensure directional solidification from the casting towards the riser located above the feed end. Besides, the pattern should be so arranged in the mould as to reduce the labour and material consumption to the minimum. The surfaces to be
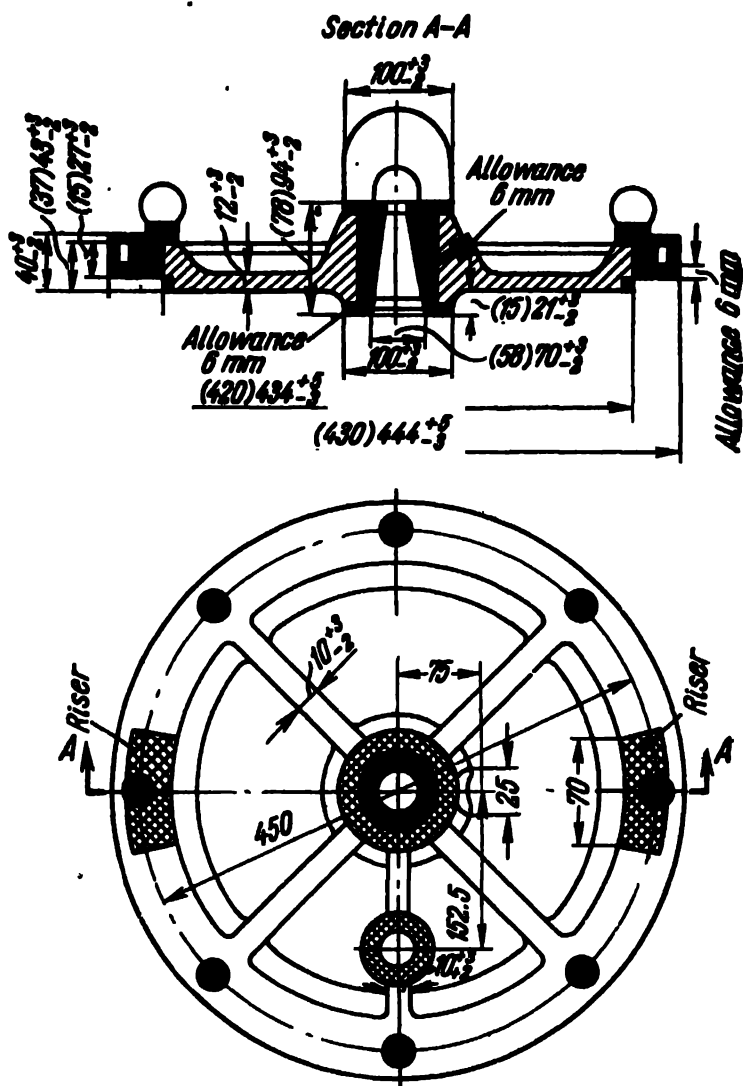
Fig. 75. Working drawing of a casting pattern

subsequently machined should be placed in the mould towards the bottom, or else vertically or in a reclined position. Large flat surfaces should never face upwards. Complex moulds should be readily accessible to inspection during assembly and provide for a reliable fastening of the cores.

Thin projections or ribs should be directed downwards or placed vertically. Thin walls of a casting are arranged vertically, in a reclined position or at the bottom to avoid short runs in pouring.

After the position of the casting in the mould has been finally decided upon, the engineer proceeds to mark off the parting planes of the mould. He notes down conventional signs (Fig. 74) on the part drawing and outlines the external contour of the casting, assigning machining allowances in conformity with the grade of accuracy required. It will be well advised to reduce
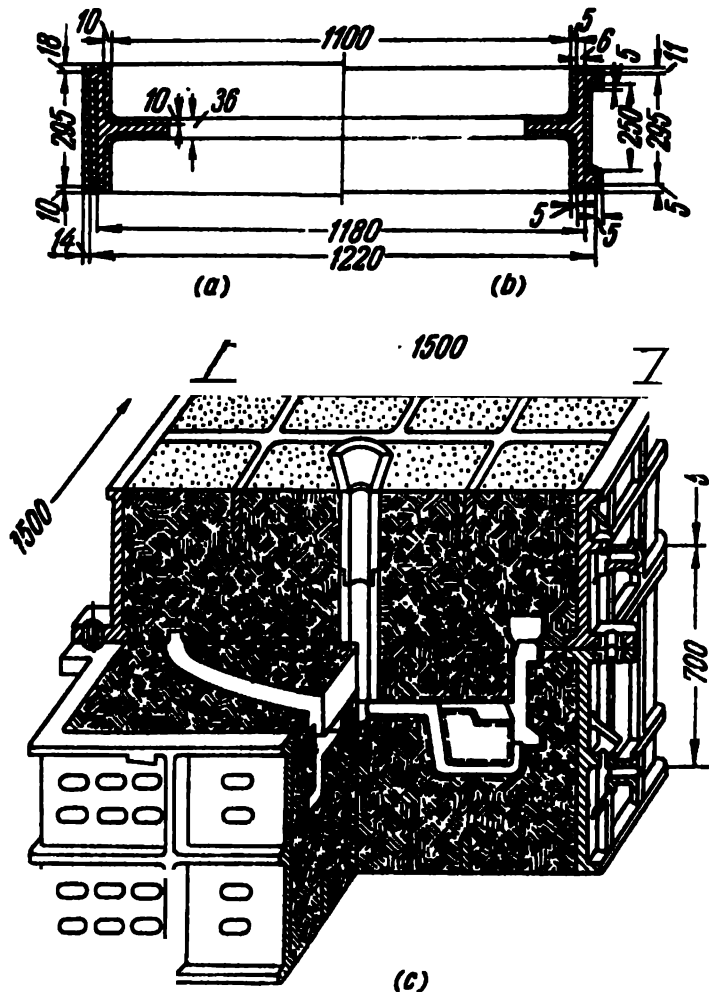
Fig. 76. Mould for a sheave made from a split pattern after reducing the machining allowances on the external surface (Vasilevsky):

a—before changing the process; b—after the process was changed; c—assembled mould

the machining allowances, or even make them negative, on those sides of the casting where the pattern has to be rapped to withdraw it from the mould in hand moulding. This will drastically reduce the weight of the casting.

Machining allowances assigned in the tractor and automotive industry are as a rule more rigid.

To ensure good directional solidification of shaped castings the drawing indicates shrinkage, wear and manufacturing allowances. Small holes are not cast. All these and other recom-

mendations are intended to obtain solid metal in the casting and greater convenience in withdrawing the pattern. Most of these recommendations tend to increase the weight of the casting or the weight of the chips to be removed.

Draft angles for castings and patterns should conform to an appropriate state standard. Draft angles are reduced or dispensed with altogether in case of machine moulding or moulding with mixtures of soluble glass.

After all the allowances and facts and figures for the pattern-marker have been transferred onto the working drawing of the casting, the pattern drawing assumes the form shown in Fig. 75. The figures in the brackets show the size of the part without

Fig. 77 Diagram of the pattern and the system for blowing the mould with carbon dioxide (Vasilevsky)
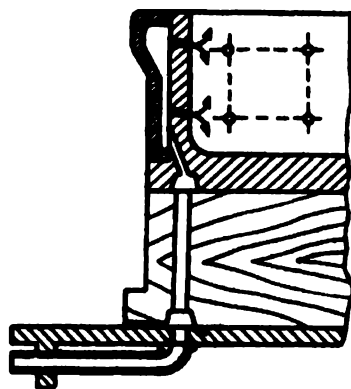
machining allowance. The size (420) 434^{+5}_{-3} denotes that after machining the diameter will be 420 mm. The allowance comes to 14 mm. The safe deviations from this size are +5 or —3 mm. After casting the diameter may vary from 431 (434 — 3) to 439 (434 + 5) mm.

All instructions of the process engineer are noted down on the working drawing with coloured china ink or coloured pencils.

The risering is marked in red, shrinkage allowances are inked in black, allowances for draft angles are denoted by black and the loose pieces of the pattern by red shading; core prints are marked with a yellow pencil and the place where the gates are joined in red.

Let us consider a few examples of the effective use of loose pieces and split patterns to remove draft angles from high vertical walls and to decrease the machining allowances. The case may be illustrated by the casting for the brake sheave of an oil drilling unit made from grade 35ХНЛ steel (Fig. 76).

The moulding box was packed with facing sand consisting of 93.8 per cent of silica sand and 6.2 per cent of refractory clay. The additions comprised 7 per cent of soluble glass with the modulus 2.6-3.0 per cent, 1.3 per cent of sodium hydrate (10-per cent solution) and 1.2 per cent of black oil. The moisture

Fig. 78. Slip-jacket moulding (Rassadin)
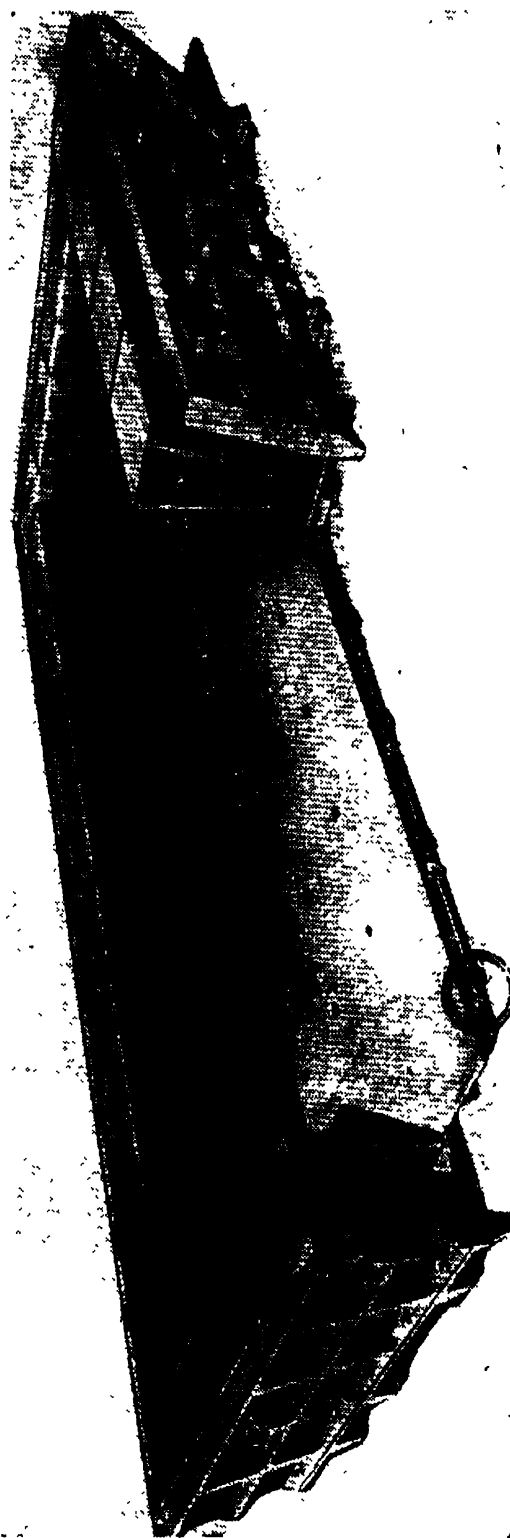
Fig. 79. Core box with replaceable parts

content was 4-5 per cent, permeability—100, green compressive strength—0.15-0.35 kg/cm² and dry tensile strength—10 kg/cm².

Fig. 77 shows a diagram of the pattern and the system feeding carbon dioxide for blowing the sheave mould.

The moulding time decreased from 17.38 to 15.13 hours, or by 12.8 per cent, and the weight of the chips was reduced from 392 to 180 kg, or by 54 per cent.

Slip-jacket moulding of large-size castings developed by M. Rassadin at the All-Union Process Designing Institute and introduced at the Elektrostal Works (U.S.S.R.) is shown in

Fig. 80. Pattern master plates for machine moulding of various castings
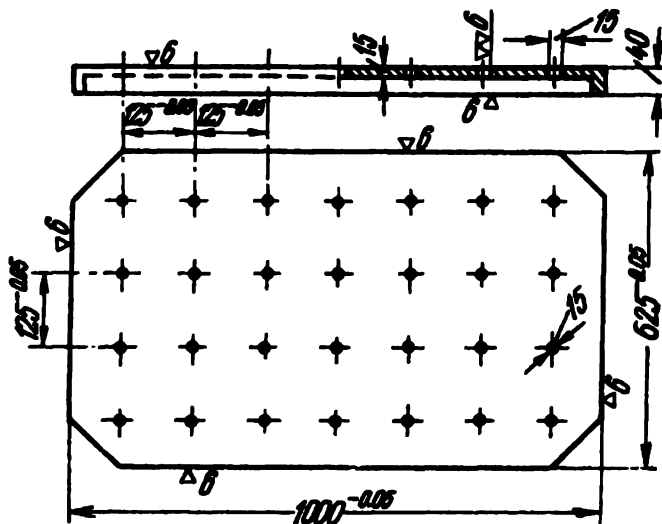
Fig. 78. Cores are machine-moulded with the aid of universal core boxes (Fig. 79). The box is composed of replaceable parts, which allow the size and shape of the cores to be changed.

The application of pattern master plates (Fig. 80) makes possible machine moulding of even single castings since the pattern can be easily installed on the plate and just as easily removed from it. It needs only several minutes to replace the pattern.

The specifications for the manufacture of pattern riggings and all the other moulding equipment are entered into process charts, which also contain all the data necessary for the pattern, core, moulding and pouring shops as well as the output rates for individual moulding operations.

17. MOULDING AND CORE SANDS FOR STEEL CASTINGS

The moulding materials and mixtures used for steel castings should be such as to ensure high quality of the cast metal and a smooth casting surface. The production of steel casting is distinguished by a high pouring temperature, high pressure head of liquid steel against the bottom and the walls of the mould

cavity, especially when the metal is poured through the bottom of the ladle, and a high chemical activity of iron and its admixtures. Iron and manganese oxides are capable of slagging the silica and produce defects on the surface of the castings.

As distinct from other sands, those used for steel castings are much more refractory and strong. The mould walls can be made chemically inert by swabbing a mould surfacing material onto the working surface or dusting it with dry materials.

The materials for moulding mixtures should be preferably obtained locally to reduce shipment costs.

Small steel castings require 4-6 m³/tons of sand and clay and large-size castings 1.5-2 m³/tons.

Let us now consider the basic rules for the selection of moulding and core sands required to obtain sound steel castings.

Cast metal may be contaminated by the particles of eroded mixture due to the three factors brought into play simultaneously—low strength of the mould walls, especially of the gating channels, improperly directed stream of liquid steel, such as might strike against the mould bottom for example, and a vortex flow of liquid metal.

The mixture tends to gain in strength as more clay or bentonite, and other bonding materials, are added. Bentonite is especially valuable since it increases the strength without requiring the moulds to be dried. This is because much less bentonite than clay is required in green sand to obtain the same strength. For example, green strength can be obtained with only 2-3 per cent bentonite, as against 4-5 per cent of clay.

An addition of clay tends to increase strength and reduce permeability. The latter is improved by drying the moulds. Although drying improves permeability it may cause the intercrystalline spaces to crack. To close up large pores on the surface, sand clay or chamotte clay moulds should be swabbed and then dried up again. This method is highly labour consuming and economically unprofitable although it has gained a broad recognition in many countries.

The use of bentonite instead of clay makes it unnecessary to dry even large moulds.

Bentonite mixtures are simple to prepare, the same methods being used as for ordinary sand clay mixtures. Requiring no auxiliary equipment, they exclude the drying and handling operations. They require neither swabbing nor dusting, which enables the production even of complex cored moulds to be automatised.

The strength of sand clay mixtures is highly promoted by additions of soluble glass. To accelerate the agglutination of the silica grains the mould or the core should be blown with carbon dioxide. These chemically solidifying mixtures possess

the same advantages as bentonite mixture but are much more stronger. Soviet foundrymen were the first to practise additions of sodium hydrate to soluble glass to promote strength. The quality of sand mixed with soluble glass is appraised from the soluble glass modulus or from the ratio $SiO_2:Na_2O$. Soluble glass with a low modulus increases the period of the mixture solidification in air and helps to obtain strong cores and shells. The modulus is decreased by adding a 10-per cent solution of NaOH. When added in small amounts (0.5-2.0 per cent) black oil also tends to increase this period.

The mixture used for large steel castings is composed of 70 per cent of silica sand, 30 per cent of used sand and, besides, 0-20 per cent of clay, 5-7 per cent of soluble glass, 1 per cent of sodium hydrate solution and 0.5 per cent of black oil. The moisture content is 3.5-4.5 per cent, green compressive strength is 0.12-0.30 kg/cm^2 and dry tensile strength 8-20 kg/cm^2 (after blowing with carbon dioxide, without drying). The ready shell of the mould or core is blown with carbon dioxide for 30 sec to 6 min, depending on their size. The application of mixtures with soluble glass is steadily gaining in scope. It improves hygienic conditions in foundry shops, reduces the amount of evolved dust, because it dispenses with drying, cuts down production costs and eliminates the causes of contamination thereby improving the quality of cast metal.

Burning-on, consisting in a spongy mass of sand adhering tightly to steel castings, is a fairly widespread phenomenon at many plants and should be avoided as much as possible. A casting with a sintered skin is very hard to clean. The cleaning of castings that go to make railway switches requires 8 to 16 manhours per ton of metal. This is the most exhausting work which may be avoided by a timely prevention of burning-on.

Theoretical studies of the causes responsible for burning-on have a long history. At present, burning-on is attributed to the three main causes—thermal, mechanical and chemical. Thermal burning-on is caused by the baking of the moulding and, especially, core sand in view of their low refractoriness and the prolonged exposure to the heat emitted by the casting. The removal of the baked layer is made more difficult by its contraction on cooling, for example in the pilger roll slot or in the hole made in a heavy casting. The metal may also penetrate the sand because it is mechanically soft and porous. This occurs in the bottom sections of high castings, given a sufficiently high pressure head. A molten metal contains many oxides, especially when manganese steel is used. Such burning-on on the castings intended for railway switches made of low-alloy steel can be prevented by adding crushed 75-per cent ferrosilicon to the mould coating.

Chemical burning-on is caused by the slagging of the surface

layer of metal with basic oxides of silicic acid present on the uncoated surface. When the slagged layer has a vitreous structure it readily separates from the casting, as was observed by A. Lyass and I. Kumanin, and its surface assumes a very clean appearance. Vitreous slags contain more acidic than basic oxides. The content of SiO_2 in soda glass varies within extremely broad limits. This feature can be very frequently utilised to obtain a clean surface on steel castings made of carbon steel.

Heavy carbon steel castings contain many basic oxides during pouring, predominantly FeO and Fe_3O_4, which combine with SiO_2 present in the mould to form a slag skin.

When casting manganese, chromium and other grades of steel, whose oxides produce no vitreous shells, the burning-on is prevented by coating the moulds with compounds which will retain a solid state at high pouring temperatures. Being chemically inert, these steels form no slag compounds with the surface oxide films.

Of late, magnesite, chromite, chrome-magnesite, dunite or forsterite (olivine with an admixture of chromite) possessing high melting points (1850-1890°C) came into broad use.

Manganese and chromium steels produce on pouring the films of complex oxides, among which basic oxides predominate; when brought in contact with the walls of a sand clay mould they yield an easily fusible crystalline slag mixture. Castings made from such steel are noticeably affected by chemical burning-on which is prevented by changing the composition of the facing sand. The facing sand for tramway spiders cast of steel Г13Л is made of baked magnesite. The mixture contains: metallurgical magnesite—18 parts by weight, baked magnesite—2 parts and clay—0.5 part; the sulphite liquor (35°Bé) comes to 0.5 per cent and water to 8-10 per cent as against the entire volume. Green compressive strength is 0.25-0.45 kg/cm², permeability—110 and moisture content 3.5-4 per cent.

Castings made of austenite heat-resistant steel, grade ЛА-1 require a mixture composed of crushed chrome-magnesite (100 parts by weight), soluble glass with the modulus 2.3 (7.5 parts), 10-per cent solution of sodium hydrate (2 parts) and water (2 parts). The green compressive strength is 0.2-0.3 kg/cm², permeability—30 and moisture content—6-7 per cent.

When refractory sands are unavailable but good refractory clays can be obtained locally instead, it is frequent practice to use crushed chamotte for heavy steel castings. The diameter of the chamotte grains varies between 4 and 5 mm. Some clay is added as a binder. Chamotte is obtained by baking the clay at 1200°C for a short time, so that its grains are slightly fused on the surface. This mixture does not fuse when brought in contact with molten steel and does not produce with oxides fusible com-

pounds which might adhere to the casting. In a thoroughly baked clay the silicates are chemically bonded and remain inert to iron oxides or react with them very slowly. When the clay is calcinated completely, its particles grow in chemical activity and the metal oxides produce spots. Weakly baked clay has a dark colouring.

Chamotte moulds are dried at 550-500°C. The moulds are swabbed with a compound consisting of chamotte flour, silica flour and graphite, and crushed coal or coke.

The facing sand is composed of fresh chamotte with the maximum grain size of 5 mm, used chamotte and refractory clay.

The moisture content of the mixture should never exceed 5 per cent.

English foundrymen make a wide use of ground chamotte, called Sheffield compo, which is a mixture containing crushed firebrick or calcinated bauxitic clay baked at 1450°C, known as grog, with a bonding addition of raw refractory clay. The crushed firebrick is ground in stone breakers and let through vibratory screens with meshes 3.45 and 1 mm. Large lumps are recrushed.

The mixture is composed of 87 per cent of ground chamotte and 13 per cent of refractory clay. The average fineness of the English sand is 1 mm which is less than the German sand (5 mm), which gives a smoother surface on the castings. The moulds are swabbed with the same compound, ground in a ball mill and containing an admixture of sulphite liquor (1 g of liquor per 12 g of water). The compound is applied to the mould immediately after the pattern has been removed. The cores are processed by the same method. The chamotte coating is covered with a layer of alundum powder from ground alundum stones rejected in the dressing shop. The compound contains: 50 kg of alundum (92 per cent Al_2O_3), 8 kg of clay and 1 kg of starch gum per 33 litres of water. Then the mould is cleaned with a soft flat brush, covered with iron sheets and dried by gas burners. The drying is conducted slowly, the temperature being raised gradually to 500°C, to prevent cracking and remove the colloidal moisture from the clay. The mould is then assembled and heated again before pouring to avoid veining. This makes it possible to fill the mould at a slow rate. The mould should be reinforced by driving springs into the face of the mould. Sometimes the moulds are calcinated up to 750-900°C.

In Czechoslovakia steel castings up to 2 tons in weight are moulded in green sand.

The foundrymen in the U.S.A. and Canada use for green-sand mould western (sodium) bentonite with the following composition: 60.89 per cent SiO_2, 22.17 per cent Al_2O_3, 3.27 per cent Fe_3O_3, 0.42 per cent FeO, 0.21 per cent FeS, 2.77 per cent MgO, 1.28 per cent CaO, 2.49 per cent Na_2O, 0.40 per cent K_2O, 0.18 per cent

TiO_2, 0.5 per cent CO_2, 0.5 per cent SO_2, 0.02 per cent the rest, and 4.78 per cent H_2O; pH=97 per cent.

In southern bentonite the adsorbed ions are chiefly calcium.

The weight of liquid metal for green-sand castings varies from 3.5 to 38 tons. Such are anvil blocks weighing up to 13 tons, anchors of ships, frames of bogies, presses and cranes, rolling and mill rollers, slag truck ladles, skids, bodies of hammers and valves, etc.

The facing mixtures are made from sand 1KPK and 1KO25A. Finer sands are used for small castings. Bentonite is added in amounts of 4-6 per cent which is dictated by the necessity to obtain green compressive strength not below 0.55-0.60 kg/cm². The mixture should be highly resistant to hot deformation. The moisture content is 3.7-6 per cent. Sand 1KPK should be mixed with 2 to 16 per cent of silica flour for better compactness and green strength.

The mixture should also contain starch gum, flour or sulphite liquor to increase the strength of the mould surfaces and slow down the erosion and crumbling of ready moulds. Adequate

Table 12

Facing Sands for Large Steel Castings Used in the U. S. A.

(Briggs)

Material and its characteristics	Mixture			
	1	2	3	4
Sand, per cent:				
old	62.8	90.7	80.6	—
reclaimed	—	—	17.8	74.4
fresh	35.2	—	—	4.5
Bentonite, per cent	1.7	2.4	1.2	4.5
Refractory clay, per cent	—	—	—	3.6
Moisture, per cent	3.2	4.8	4.3	4.5
Flour, per cent	0.3	1.0	0.4	0.5
Silica flour, per cent	—	5.9	—	12.5
Fineness, AFS	51	46	49	50
Green compressive strength, kg/cm²	0.47	0.68	0.53	0.69
Hot strength, kg/cm² at °C				
260	5.8	—	—	—
537	7.1	19.0	—	—
818	29.1	61.0	—	25 0
1093	7.9	33.0	—	53.0
1378	0.3	3.6	2.1	0.29
Contraction at 818° C, cm	—	—	—	0.0002
Expansion at 818° C, cm/cm	—	—	—	0.015

strength is obtained by "seasoning." the moulds in air or by drying their surfaces. Wood flour, iron ore or pitch must be added to increase the pliability.

The permeability of sand ranges between 70 and 250, but not below 70.

The typical compositions and properties of mixtures are illustrated in Table 12.

The most essential difficulty in green sand moulds accrues from the fact that the sand should be rammed in very tightly for, otherwise, it will lose some of its erosion resistance which will spoil the surface of the casting. For this reason the ramming should be done with a sandslinger. Another difficulty arises from the sands adhering to the patterns. Besides, the moulds have to be assembled and filled immediately, which likewise involves certain difficulties.

For all that, green-sand castings are superior in quality to dry-sand castings. Green-sand moulding requires less time, the accuracy of the castings is higher, hot cracking is not so intensive and the costs of production are lower.

18. GATING SYSTEMS FOR STEEL CASTINGS

Poor design and wrong size of the gating system as well as improper temperature and pouring rate account for nearly seventy-five per cent of rejects incurred in steel foundries.

Attempts to arrive at a general theoretical solution or a calculation formula have so far been futile. For this reason, preference is usually given to one or two decisive factors, while disregarding the others. Experienced foundrymen are guided in their work by the empirical data provided by long years of production experience. Although the amount of rejects caused by the wrong design of the gating system is being steadily reduced, the consumption of extra liquid metal to combat this defect is still very high.

A rational design of the gating system elements results in good mechanical properties of the ready product, improves its appearance and raises the economic indices of production.

In calculating the dimensions and selecting the design of a gating system, at least three demands must be complied with:

(1) suitable design of the gating system to ensure unfailing operation of all risers;

(2) adequate temperature of steel in pouring to prevent the steel from freezing at the head or in the middle of the stream, nor in the remotest sections of the casting, before pouring is completed;

(3) properly selected optimum pouring rate to avoid the formation of oxide films and veining or the collapse of the mould

top which usually occur at slow pouring rates and, conversely, to avoid cracking and contamination of the metal with sand as well as the entrapment of the air and gases by metal at too fast a pouring or when the taper of the sprue opening and the dimensions of the other elements of the gating system are assigned incorrectly.

The designs of the gating channels depend on the pouring practice and are provided in several standard versions (Fig. 81).
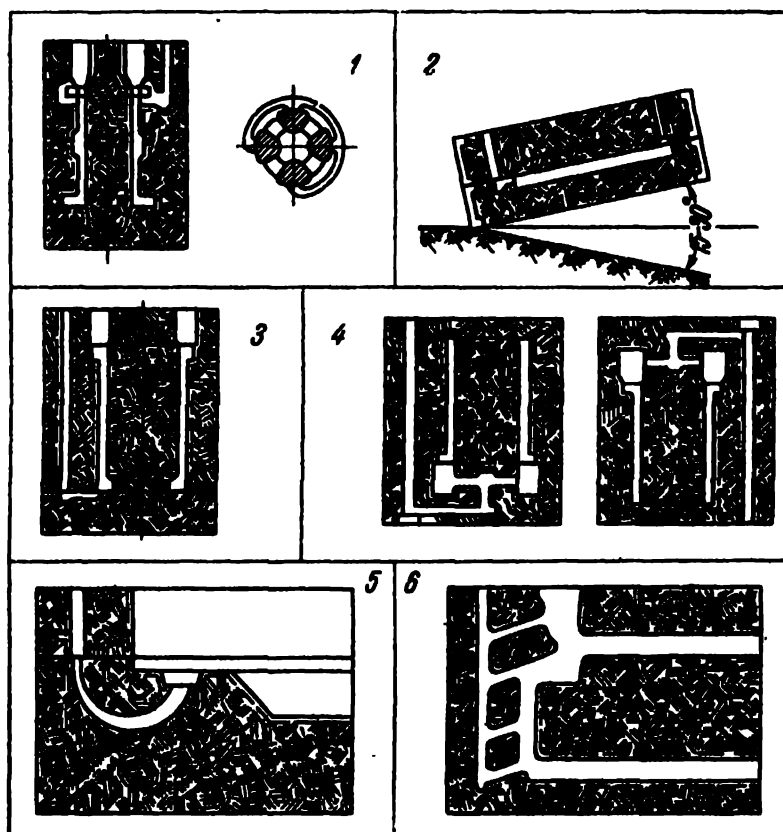
Fig. 81. Standard designs of gating systems for steel castings:

1—top gate; *2*—bottom gate with partial reversal; *3*—side gate; *4*—bottom gate with complete reversal; *5*—horn gate; *6*—step gate

A top gate is used for broad and low castings. In the base of a vertical sprue the flow of metal bifurcates and passes first to the mould through an annular screen in broad band-like streams and then to four risers. The second design is employed for castings poured in a horizontal or tilted position. A flow-off is provided at the highest point of the mould cavity. To increase the pressure of metal during contraction the slant of the mould is reversed about 30° after pouring and the riser is brought to a position above the rest of the piece. This partial reversal method has found a broad application in the U.S.A., Czechoslovakia and some other countries.

A bottom gate has the disadvantage of allowing the metal to fill the riser in a somewhat cooled condition, after it has lost a considerable part of its superheat, thereby producing a negative temperature gradient (see Fig. 44). If the metal enters the mould cavity from below as shown in the fourth diagram it will pass almost completely through the risers located at the bottom. After a solid skin has formed on the casting the mould is turned upside down. The risers will be brought to the top and will feed the casting at a positive temperature gradient. This is the complete reversal method. A horn gate (Fig. 81, 5) is a variety of bottom gate. One objection to its use is a tendency for producing a fountain effect in the casting. As a remedy, a reverse horn gate may be used with the large end next to the casting and the small end connected to the sprue. Generally, any gating system with fanning out channels is called a subnormal pressure system.

A step gate is intended for large-size castings (Fig. 81, 6). Hot metal fills the riser through the upper gate at the end of pouring (see also Fig. 48).

Sometimes the castings are poured through a vertical knife gate. This system can be likened to a step gate with an infinitely large number of gates.

Each of the above standard designs has its own merits and demerits. The top gate is used, for example, for investment pattern castings. It has the advantage of a short route from the pouring cup to the mould cavity. But the eroded particles of the channel walls, especially in the sprue base, rise to meet the stream of steel and get entrapped in the casting as nonmetallic inclusions. No dirt traps will be of any avail, since they will increase the flow of liquid metal and disturb directional solidification. A common sprue for many small castings may serve as a riser.

The side gate is used for large-size and long castings, such as tubes and hollow cylinders. The riser is arranged near the runner. Pencil flow-offs provided at the top preclude the formation of an air pocket. The flow-offs pass the air, but not the metal, since they have a very small cross section and the steel fluidity is nearly zero. To increase the metal head on contraction the tilt of the mould is reversed 15-30° after pouring and the riser is brought to a position above the rest of the piece. The additional operation of reversing the mould position after it has been filled constitutes the shortcoming of this system.

The side gate without reversal is frequently applied for thin-walled castings, such as frames, plates, brackets, etc. Individual castings are poured in a tilted position. The surface quality improves, if the pouring is done as shown in Fig. 81, 2.

The system shown in diagram 3 has gained a wire recognition. It ensures an adequate surface quality since the dirt is carried by the stream upwards and into the riser. However, directional

solidification takes an opposite course (see Fig. 44). The sixth design partly compensates for the shortcomings of the third design, because hot metal enters the riser through the upper gate at the end of pouring. The knife gate (see Fig. 21) takes an intermediate position between the third and sixth designs. The fourth design ensures a perfect directional solidification but involves the reversal of the mould.

The horn gate produces a good surface on a casting and in no way differs from the third design, possessing at the same time all its shortcomings. High pouring rates may cause the entrapment of air and the casting will be affected by gas holes. This shortcoming is inherent in all gating systems with fanning out channels in which the motion of the metal stream builds up a hydraulic vacuum attended by sucking in of gases from the pores of the mould facing material.

In addition to these six designs of gating systems, other systems may be encountered in foundries as well.

The design of a gating system should be appraised not only from the very widespread hydraulic standpoint but also from the thermal features of feeding.

The second demand (see p. 141) requires that the remotest sections of the casting be filled before solidification sets in. Obviously, the pouring time should satisfy the inequality

$$\tau_1 \leqslant \tau_2, \tag{88}$$

where τ_1 = time of cooling during pouring without riser;

τ_2 = solidification time, from the pouring temperature θ_p to the liquidus temperature, θ_l, i. e., the time for the loss of steel superheat.

The time τ_1, and, hence, the volumetric and weight pouring rates, depend, therefore, on the thermophysical properties of the metal and mould, in other words, on the mould heat transfer and the degree of the metal superheat.

Today, several methods can be used to determine the time during which the metal is cooled in the mould to the liquidus temperature.

We shall consider one of these methods developed by Charles Trenklé.

According to G. Schwarz and N. Chvorinov, the cooling time of plates and balls from the pouring temperature θ_p to the liquidus temperature θ_l can be expressed as dependent on the square of the referred thickness of the casting:

$$\tau_2 = (AR)^2 \tag{89}$$

for the plates and

$$\tau_2 = (BR)^2 \tag{90}$$

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for the balls, where

$$A = \pi \left[\frac{\gamma_1' \, c_1' \, (\theta_p - \theta_l)}{2\Delta\theta\gamma_2 c_2\theta_l} \right] \qquad (91)$$

and

$$B = \left(\frac{3}{\Delta\theta \sqrt{\pi}} \sqrt{1 + \frac{\pi\gamma_1' \, c_1'(\theta_p - \theta_l)}{3\gamma_2 c_2\theta_l}} \right) - 1. \qquad (92)$$

The ratio between the coefficients is the most important factor in shaping a ball casting; we shall refer to it as the geometrical factor:

$$\Gamma = \frac{B}{A} = B \left(1 = \frac{\sqrt{\pi}}{6} \Delta\theta B \right). \qquad (93)$$

For the cylinders $\Gamma = 0.95$.

The thermophysical coefficient A depends on the quality of the alloy and the amount of superheat in pouring. For dry-sand moulding this coefficient should be increased by 25 per cent and for loam or cement moulding by 35 per cent.
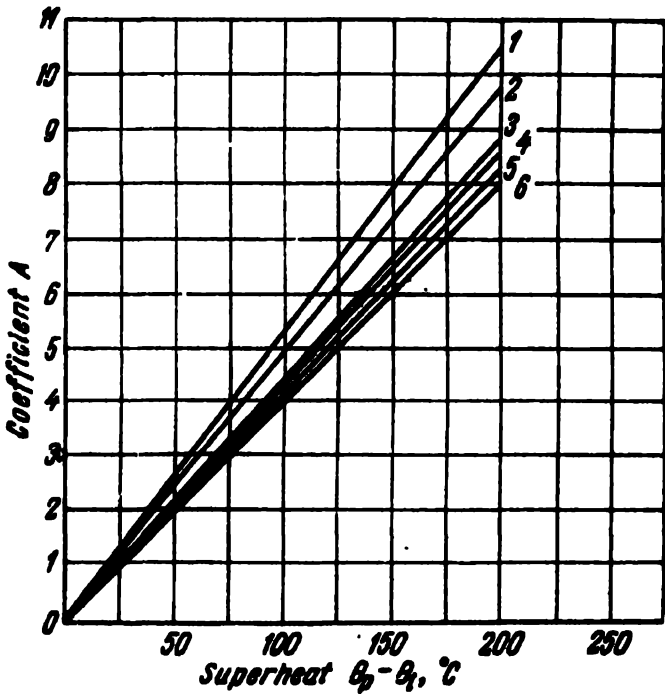
Fig. 82. Alignment chart for determining the coefficient A in Eq. (93) depending on the metal superheat in pouring (Charles Trenklé):

1—eutectic pig iron; *2*—hypoeutectic pig iron; *3*—pure copper; *4*—pure aluminium; *5*—steel with 0.3 per cent carbon; *6*—pure iron

For plates cast in green-sand moulds the coefficient A can be found from the alignment chart in Fig. 82.

The cooling time for liquid metals and alloys is indicated in Table 13.

Table 13

Cooling of Liquid Metals and Alloys from the Pouring Tempe ature to the Liquidus Temperature
(Plates and Balls in Green Sand Moulds)
(Charles Trenklé)

Parameter	Pure Iron	Steel 0.3% C	Pig Iron	
			hypoeutec-tic*	eutectic
I. Cooling in green-sand mould at θ_2—0				
Pouring temperature θ_p on attaining θ_l on interface, °C	1689	1667	1377	1267
Liquidus temperature θ_l, °C. . .	1535	1515	1252	1152
Solidus temperature θ_s, °C	1535	1485	1152	1152
(a) Pouring temperature at minimum superheat				
Pouring temperature θ_p, °C . . .	1650	1590	1370	1260
Superheat $\theta_p - \theta_l$, °C	115	75	118	108
Theoretical temperature θ_0 on interface, °C	1500	1445	1245	1145
Difference $\theta_0 - \theta_l$, °C	—35	—70	—7	—7
Coefficient A for plates	4.94	3.25	5.73	5.71
Cooling time τ from θ_l to θ_p . .	6.1	2.64	8.2	8.15
Geometrical factor $\Gamma = \dfrac{B}{A}$ for ball	0.912	0.924	0.905	0.900
Cooling time τ of balls from θ_p to θ_l, sec	0.56	0.25	0.75	0.73
(b) Pouring temperature at 12.8% superheat from θ_l				
Pouring temperature θ_p, °C	1730	1700	1410	1300
Superheat $\theta_p - \theta_l$, °C	195	185	158	148
Theoretical temperature θ_0 on interface, °C	1570	1546	1280	1180
Difference $\theta_0 - \theta_l$, °C	+35	+30	+28	+28
Coefficient A for plates	8.37	8.02	7.67	7.82
Cooling time τ of plates from θ_p to θ_l, sec	17.5	16.1	14.7	15.3
Geometrical factor $\Gamma = \dfrac{B}{A}$ for ball	0.875	0.885	0.890	0.880
Cooling time τ of ball from θ_p to θ_l	1.49	1 40	1.30	1.32

δ—plate thickness; d—ball diameter

II. Cooling in dry sand-mould at $\theta_2 = 0$ *and* $\theta_0 = \theta_l$

$A_{dry\ plate} = 1.25 A_{green\ plate}$; $B_{dry\ ball} = 1.25 A_{green\ ball}$; $\Gamma_{dry\ ball} = \Gamma_{green\ ball}$; $\tau_{dry\ plate} = 1.56 \tau_{green\ plate}$; $\tau_{dry\ ball} = 1.56 \tau_{green\ ball}$.

III. Cooling in cement or clay mould

$A_{cement\ plate} = 1.35_{green\ plate}$.

* Chemical composition of pig iron: 3.04 per cent C. 1 39 per cent Si and 0.53 per cent Mn.

It follows from the data in Table 13 that, even at a very small superheat, the cooling time from θ_p to θ_t increases in proportion to the square of superheat. In case of different alloys and metals 'the' geometrical factor for the balls remains practically the same, $\Gamma = \dfrac{B}{A} = 0.9 \pm 2.8$ per cent, but it must not be ignored when calculating the cooling time of the balls, proportional to the square of Γ.

In calculating the cooling time of the castings the thermophysical constants were borrowed from Table 14 compiled from different sources.

Table 14

Thermophysical Constants of Pig Iron and Steel

(Charles Trenklé)

Metal	Specific weight, kg/dm³		Heat conductivity λ_i', kcal/cm	Heat capacity c_i', kcal	Temperature conductivity b_i', cal
	liquid state	solid state			
Pure iron	6.90	7.85	0.03	0.19	0.152
Grade 30Л steel	7.00	7.80	0.03	0.19	0.150
Pig iron:					
hypoeutectic	7.10	7.25	0.03	0.19	0.148
eutectic	7.25	7 25	0.03	0.19	0.147

The thermophysical constants of moulding sands are illustrated in Table 15.

Table 15

Thermophysical Constants of Moulding Sands

Moulding sand	Temperature θ, °C	Specific weight when packed, γ_2, kg/dm³	Heat capacity, c_2, kcal/kg	Heat conductivity, λ_2, kcal/cm
Green sand with 9% clay	0	1.55	0.27	5-5.7
Sand with 4% moisture	100	—	0.19	1.1
Sand with 9% clay	1000	—	0.21	1.24
Dry sand	1000	1.50	0.19	0.90

As in the previous calculations of the volume of the solidified metal, the solidification time from θ_p to θ_l may be determined by the following equation which holds for all cases

$$\tau = (A\Gamma R)^2, \tag{94}$$

where Γ = geometrical factor of the mould;
A = thermophysical coefficient;
R = referred thickness of the casting wall, m.

The coefficient A can be found from the chart in Fig. 82 and the geometrical factor Γ from Fig. 83, as a function of the length-to-thickness, length-to-width or length-to-diameter ratios of the casting.
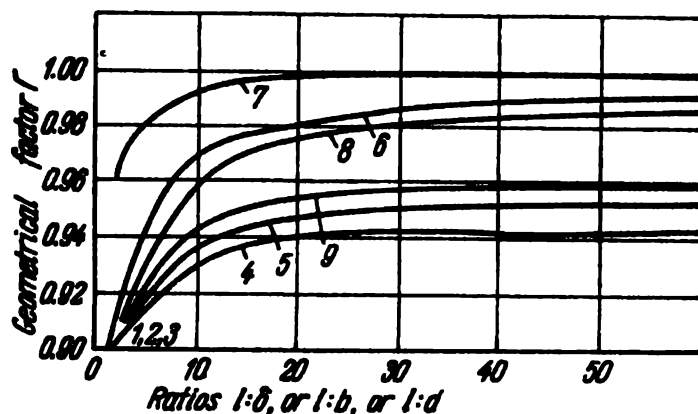
Fig 83. Change in the geometrical factor Γ as a function of the casting dimensions:

1-3—balls, cubes and cylinders ($\Gamma = 0.90$); 4—cylinders ($l:d$); 5—square bars ($l:\delta$); 6—square plates ($l:b$); 7—rectangular plates ($l:\tau$); 8—rectangular plates ($b=1:2$; $l:\delta$); 9—bars ($l:b$)

In the case of solid bodies and at a minimum superheat of grade 30 Л steel by 5 per cent, the temperature $\theta_p = 1590°$ C and $\theta_l = 1515°$ C. Then, $\theta_p - \theta_l = 75°$ C.

The coefficient A, equal to 3.15, will be found on the alignment chart (Fig. 82).

The cooling time necessary to obtain the liquidus temperature at the minimum value of the geometrical factor Γ can be found from the following equations:

$$\tau = \left(1 \times 3.15 \,\frac{\delta}{2}\right)^2 = 2.47\delta^2 \tag{95}$$

for the plate,

$$\tau = \left(0.95 \times 3.15\frac{d}{4}\right)^2 = 0.56\,d^2 \tag{96}$$

for the cylinder,

$$\tau = \left(0.90 \times 3.15\frac{d}{6}\right)^2 = 0.222\,d^2 \tag{97}$$

for the ball.

To determine the cooling time of hollow bodies, use can be made of the corrections for the coefficient Γ, provided by A. Veinik, for example—instead of Γ, to use $\Gamma = 1 + \frac{1}{2}\delta$ for a hollow cylinder cooled from inside and $\Gamma = 1 + \delta + \frac{1}{8}\delta^2$ for a hollow ball. Diepschlag or Czikel corrections may also be used. When steel containing 0.3 per cent carbon is pured at 1590° C the cooling time of a hollow casting can be found from the equation

$$\tau = (\Gamma R - A)^2, \qquad (98)$$

where Γ = geometrical factor;

R = ratio between the volume of the hollow casting and the external surface;

A = coefficient to be found on the alignment chart (Fig. 83).

The above equations give good approximate results and can be useful in the selection of temperature and pouring rate. The time τ found from Eqs (94) through (98) should obviously depend on the nature of the gating system or, in other words, on whether the metal is fed from the top, from the side or from the bottom. When top gating is employed at θ_p, the temperature of the metal at the mould bottom may drop to θ_t on the interface. As the mould is being filled the temperature of the metal at the highest level will be θ_p. Deeper in the casting, the metal cools slowly down to θ_t. The temperature distribution on the side surfaces is rather uniform. Faster rates of cooling are observed at the bottom of the casting. But it is hardly possible to express the pouring rate, for example, by means of Eq. (93):

$$v_p = \frac{h}{\tau}, \qquad (99)$$

where h = height of the casting, m;

τ = pouring time found from Eqs (95) through (99), sec.

The degree of mixing of the metal poured from the top, the convection and the cooling rates in the upper and bottom areas of the casting are unknown when we determine the cooling of the top and bottom of the casting.

The calculations grow more involved when the cooling time of a shaped, variable section casting has to be determined.

The correction factor K is introduced into Eq. (94) to obtain more or less reliable cooling time

$$\tau = K(A\Gamma R)^2. \qquad (100)$$

For top gating K is assumed equal to 0.92, for bottom gating to 0.87 and for side casting of thin plates to 0.85.

Practical application of these approximate calculations in many countries produced very good results.

Soviet foundrymen use the empirical formula developed by K. Sobolev

$$\tau = S \sqrt[3]{\delta G},\qquad(101)$$

where τ = pouring time, sec;

S = experimental coefficient, found for steel castings from Table 16, compiled by G. Dubitsky from experimental data;

δ = mean thickness of the casting wall, mm;

G = weight of the casting, gating and risering, kg.

Experimental Values of the Coefficient S for Steel Castings

(Dubitsky)

Pouring temperature and fluidity	Bottom gating	Side gating	Top gating
Normal	1.3	1.4	1.5-1.6
Increased	1.4-1.5	1.5-1.6	1.6-1.8

The time needed to cast a plate $8 \times 128 \times 250$ mm in size and weighing about 25 kg with the riser and gates is, according to Eq. (101), equal to

$$\tau = 1.55 \sqrt[3]{8 \times 25} \approx 9 \text{ sec}$$

and, according to Eqs (89) and (94), to

$$\tau = 0.85 \left(1 \times 7.6 \frac{0.8}{2}\right)^2 \approx 2.6 \text{ sec,}$$

i. e., almost 3.5 times more. A thin plate poured in too slowly may develop veining. It follows from this that the filling rate may be calculated from the cooling time only in individual specific cases.

A plate $50 \times 1,000 \times 2,000$ mm in size should be served by 4 risers, amounting to 0.21 of the weight of the casting. A quarter of the plate will weigh 195 kg and the riser $0.21 \times 195 = 41$ kg. The approximate total weight of the metal poured into the mould will be 1,000 kg.

According to N. Sobolev and G. Dubitsky, the pouring time is

$$\tau = 1.4 \sqrt[3]{50 \times 1,000} = 52 \text{ sec.}$$

Using the Trenklé method, with the correction factor of 0.85,

$$\tau = 0.85 \times 2.47 \times 5^2 = 62.5 \text{ sec.}$$

In both cases, the pouring time was almost the same.

The choice of the pouring temperature for steel castings also depends on steel fluidity, or on steel composition and superheat.

V. Gruzin has calculated that the maximum temperatures attained in steel smelting furnaces without fusing the roof amount to 1640° C in an open-hearth furnace with a Dinas roof, to 1680° C for a furnace with a chrome-magnesite roof, provided the air is enriched with oxygen, to 1740° C in a converter, to 1690° C in an electric arc furnace with a Dinas roof, to 1850° C in the same furnace with a chrome-magnesite roof with oxygen refinement, and to 1850° C in an induction furnace. However, the temperature of steel need not be raised necessarily to this point in actual practice.

In a ladle of 15-ton capacity the steel cools under slag at a rate of 4 or 6°C per minute which depends on the temperature the ladle refractory lining is raised to before tapping. When the steel is spout-poured at usual rates the temperature drops by 40-60°C. To prevent skull formation in the ladle bottom, the temperature of the steel should never be allowed to drop below 1590° C at the end of pouring.

After reaching its maximum at a high superheat the steel fluidity hardly changes at all afterwards and it is therefore unpracticable to raise the pouring temperature above this critical point.

Definite steel pouring temperatures have been established at Soviet foundries depending on the pouring method. For reference, use Table 17.

The following conclusion can be drawn from the data in Table 17. The heavier are the parts and the smaller their dimensions, the lower should be the pouring temperature. Although shrinkage cavities grow smaller in size at low pouring temperatures, the metal around them tends to become more porous. If alloy, less fluid, steels are used the casting may be affected by undersurface gas holes and nonmetallic inclusions.

Safe limits of pouring temperatures are given on the diagrams in Fig. 84 compiled by V. Gruzin and V. Saveiko from the experimental data on three most widespread grades of steel and for castings with different wall thickness (shaded portions). The diagrams show that each referred wall thickness and each grade of steel are distinguished by their own limit of optimal pouring temperatures. Disregard for these limits may cause defective castings. It can be assumed that many grades of steel accepted for shaped steel castings possess poor castability and require that their chemical composition be corrected to promote their soundness. The search for such steels with a good castability is continued even today. It can be noted, however, that the generally

Table 17

Safe Pouring Temperatures for Some Machine Elements
(Gruzin)

Product	Weight or size	Grade of steel	Temperature θ, °C		Per cent of super-heat over θ_l
			actual	uncorrected value of optical pyro-meter	
Parts for the body of steam turbine cylinders	Thickness 22-30 mm	25Л	1565-1590	1450-1470	2.8-4.5
Cylinder bodies of steam and gas turbines	Thickness 30-45 mm	25Л heat resis-tant (15% Cr, 15% Ni)	1540-1560	1400-1510	3.5-4.0
Thin-walled flanges, rings, etc.	Diameter below 300 mm	25Л	1545-1580	1420-1450	2.2-3.8
Flywheels, gears, runner rollers, driving wheels, sprockets	Diameter below 1 m	35Л	1585-1570	1410-1440	1.0-3.3
Sheaves	Diameter below 1 m	25Л	1545-1565	1420-1440	1.5-2.8
Frames for earth diggers, crane and car trucks and moulding boxes	—	35Л	1535-1555	1410-1430	1.0-2.2
Rubbing parts of earth diggers, bulldozers, plates of stone cru-shers and mills, wheel tyres, ca-terpillar tracks	Weight below 500 kg	Г13Л	1405-1435	1350-1380	1.0-2.0

recognised grades 20Л-30Л are undoubtedly the best among the carbon steel category.

The size of the elements of gating systems is determined on the basis of Bernoulli's equation. Let us take one of the simplest versions of this equation and consider the change in the total

Predominant thickness of casting, wall, mm

150 125 100 75 50 25 15

25 Л

Fluidity

Linear shrinkage

Concentration of volumetric shrinkage

Resistance to hot cracking

Evolution of gas

Burning-on

Mechanical properties

25 ХМ

Fluidity

Linear shrinkage

Concentration of volumetric shrinkage

Resistance to hot cracking

Evolution of gas

Burning-on

Mechanical properties

25 ХМФ

Fluidity

Linear shrinkage

Concentration of volumetric shrinkage

Resistance to hot cracking

Evolution of gas

Burning-on

Mechanical properties

1450 1500 1510 1520 1530 1540 1550 1560 1570 1580 1590 1600 1700 1800

Actual pouring temperature, °C

Useful effect of steel superheat

Unexplored area

Fig. 84. Alignment chart for pouring temperatures versus the predominant thickness of the casting wall (Gruzin, Saveiko)

Fig. 85. Change in the total energy of 1 kg of liquid flowing through an inclined pipe (Bernoulli)
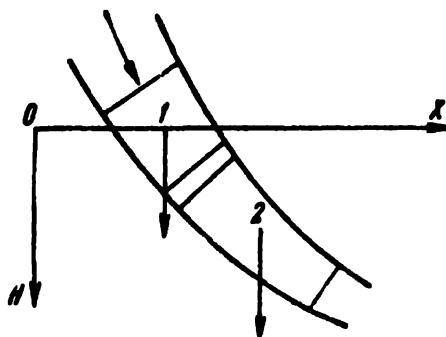
energy of 1 kg of liquid flowing from position *1* to position *2* (Fig. 85)

$$\left(\frac{v_1^2}{2g} + \frac{P_1}{\gamma} + h_1\right) - \left(\frac{v_2^2}{2g} + \frac{P_2}{\gamma} + h_2\right) = H_l, \qquad (102)$$

where $\frac{v_1^2}{2g}$ and $\frac{v_2^2}{2g}$ is the kinetic energy of 1 kg of liquid in positions *1* and *2*;

$\frac{P_1}{\gamma}$ and $\frac{P_2}{\gamma}$ is the piezometric energy in positions *1* and *2*;

h_1 and h_2 is the potential energy in positions *1* and *2*;

H_l = energy lost during movement from position *1* to position *2*.

Let us assume that the origin of our ordinate is in the centre of gravity of a certain volume of liquid in position *1* and that the liquid is at rest (initial level of steel in a stopper ladle). Then, the linear velocity of movement $v=0$, and the term $\frac{v_1^2}{2g}$ will become zero.

Let us further assume that the flow is not separated into individual streams and that the external pressure P_1 exerted on the level of liquid and the pressure P_2 in position *2* are the same with the liquid in continuous flow. Then, the values $\frac{P_1}{\gamma}$ and $\frac{P_2}{\gamma}$ may be neglected. Originally, the level $h_1=0$. After transformations we shall obtain Toricelli's equation for an ideal liquid which is not characterised by viscosity

$$v = \sqrt{2gH}, \qquad (103)$$

where g = acceleration due to gravity;
H = height of lost head.

Liquid metal obeys Newton's laws as long as a liquid phase predominates in it. For this reason, the motion of the molten metal through the gating system channels is governed by the laws of hydraulics and Eq. (103) can therefore be applied to the calculations given below if account is taken of the friction between particles in turbulent flow (turnings, expansion and compression of the stream and friction against the channel walls). After introducing this correction, Eq. (103) becomes Bernoulli's equation

$$v = \mu \sqrt{2gH}. \qquad (104)$$

The calculations of the stream movement along a multichannel gating system grow in complexity since the effect of the piezometric energy can no longer be disregarded.

Multiplying both sides of the equation by f and γ, where f is the cross section of the flow channel and γ is the specific weight of the liquid, we shall obtain the mass or weight flow rate per unit time

$$v_w = f\gamma\mu\sqrt{2gH}. \tag{105}$$

Since the weight flow rate is equal to the weight of the poured metal G, divided by the pouring time τ, then solving Eq. (105) with respect to f we obtain

$$f = \frac{G}{\tau\gamma\mu\sqrt{gH}} \text{ sq cm.} \tag{106}$$

Since for liquid steel $\gamma = 7.0 \text{ kg/dm}^3$, then $\gamma\sqrt{2g}\,0.001 = 0.31$ and

$$f = \frac{G}{0.31\,\tau\mu\sqrt{H}} \text{ sq cm.} \tag{107}$$

This equation is known as Osann's equation.

Let us consider two gating systems (Fig. 86) operating under pressure and under subnormal pressure conditions.

The cross sections of channels in a pressure system with lip pouring (Fig. 86, 1) is determined by the constricted spot MN where the sprue and the gate meet. The sprue is filled with metal to the level AB. Thus, the area f in Eq. (105) refers to section MN and the head H_m should be referred to the difference between the levels from the section AB to the lowermost point of the section MN. When the metal is poured through a pouring basin (Fig. 86, 2) the head H_m should be counted from the level of metal in the basin to the section MN. In pouring through the ladle bottom the head H_m should be taken between the level of steel in the ladle to the section MN (Fig. 86, 3). The flow will be determined not by the section MN but by the hole in the bottom-pour ladle.

Steel may be poured from a bottom-pour ladle with the stopper lifted to a full or partial opening. The first method is more preferable because the steel enters the mould without spattering and the steel oxidises in air not so intensively. But when a large number of identical moulds are to be filled the stopper has to be partially shut off at the beginning of pouring, when the ladle is still full, to create restriction to flow.

Let us consider the calculation of the gating system elements for these two pouring methods.

The first method. The taphole in the ladle bottom is opened up and nothing restricts the flow of metal. The pouring time is determined by the formula of N. Sobolev and G. Dubitsky (105) and the coefficient S is selected from Table 18. The duration of pouring depends on the given diameter of the stopper taphole,

the mean pressure head of the steel and the weight of the steel poured into the given mould and is determined by the equation

$$v_w = \frac{G}{\tau} = \mu_l F_t \gamma_l \sqrt{2gH_m} \,, \qquad (108)$$

where G = weight of the metal poured into the mould, kg;
τ = pouring time, sec;
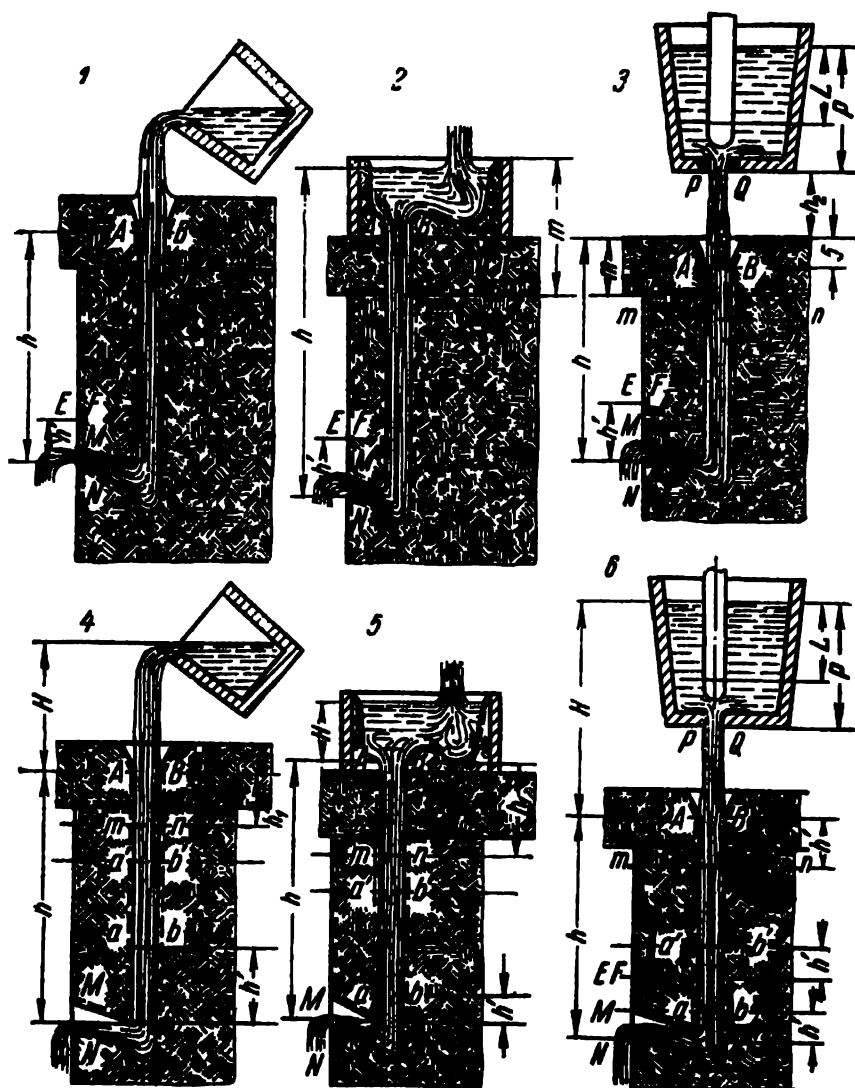
Fig. 86. Pressure (1, 2 and 3) and subnormal pressure (4, 5, 6) gating systems. Lip pouring (1, 4), basin pouring (2, 5), bottom pouring (3, 6) (Henon)

μ_l = coefficient of steel flow from the ladle taphole, assumed equal to 0.8;
F_t = cross-section of the stopper taphole, sq dm;
g = acceleration due to gravity, equal to 98.1 dm/sec^2;
H_m = mean head in the ladle in filling the given mould, dm;
γ_l = specific weight of liquid steel, equal to 7.0 kg/dm^3.

If the cross section of the stopper taphole is to be selected, the mean head of steel in the ladle H_m must be specified first and, vice versa, the mean head has to be found if the diameter of the taphole is known.

The mean head during pouring of all steel from a full ladle can be found from Eq. (109) (see Fig. 86, 3):

$$\sqrt{H_m} = \frac{\sqrt{P} + \sqrt{P-L}}{2}, \tag{109}$$

where P = level of steel in the ladle at the start of pouring, m;
L = level of steel in the ladle at the end of pouring, m.

The following formulas can be used to calculate the cross sections of gating channels:

$$F_{max} = F_t \frac{\mu_t \sqrt{P}}{\mu \sqrt{h_i}} \tag{110}$$

and

$$F_{min} = F_t \frac{\mu_t \sqrt{P-L}}{\mu \sqrt{h_f}}, \tag{111}$$

where μ = flow rate coefficient of the gating system;
h_i = initial head of the gating system, cm;
h_f = final head of the gating system, cm.

The largest result should be taken from the two solutions of Eqs (110) and (111). Two methods are possible in this case.

The first method disregards the effect produced by the energy of the stream of the flow rate in the gating system. For Eqs (110) and (111) use is made of the initial head in the gating system

$$h_i = h - 5 \text{ cm}, \tag{112}$$
$$h_f = m - 5 \text{ cm}. \tag{113}$$

The second method takes into account the effect of the falling stream energy on the flow rate in the gating system.

The pressure head is found from the formulas:

$$h_i = P + h_s + h - (h_i' + h_i'') \tag{114}$$

for the initial moment and

$$h_f = P - L + h_s + (m-5) - (h_f' + h_f'') \tag{115}$$

for the final moment;

where h_i' and h_f' = loss of head as the metal is poured from the ladle taphole at the start and end of pouring;
h_i'' and h_f'' = loss of head for impact in the plane AB for vortex motion in the sprue and for other resistances not accounted for by the flow rate coefficient of the gating system.

The loss of head during the pouring of metal through the tap-hole is calculated in conformity with hydraulics formulas

$$h'_l = (1 - \varphi^2) P \quad \text{and} \quad h'_f = (1 - \varphi^2)(P - L); \tag{116}$$

$$h''_l = K \frac{(v'_1 - v'_2)^2}{2g} \quad \text{and} \quad h''_f = K \frac{(v''_1 - v''_2)^2}{2g}, \tag{117}$$

where φ = velocity coefficient of metal flow from the taphole which may be taken as approximately equal to the coefficient of flow rate from the ladle $\mu_l - \varphi \approx \mu_l = 0.8$;

v_1 and v''_1 = mean velocities of particles of the falling stream of metal at the level AB at the initial and final moments, m/sec;

v_2 and v''_2 = mean velocities of the stream particles in the sprue at the level AB at the initial and final moments, m/sec.

Besides,

$$v'_1 = \sqrt{2g[P + h_2 + h - (h + 5) - h'_l]} \tag{118}$$

and

$$v''_1 = \sqrt{2g(P + h_2 + m)}. \tag{119}$$

The notation in brackets under the radical shows the total heads at the initial and final moments of pouring, provided the lowest level of metal at the constricted spot is taken for the initial moment and the topmost level of metal in the cavity of the filled mould for the final moment.

The velocities in the sprue at the level AB at the initial and final moments are

$$v'_2 = \frac{Q'}{F_{sp}} \quad \text{and} \quad v''_2 = \frac{Q''}{F_{sp}}, \tag{120}$$

where Q' and Q'' = flow rates of metal from the ladle at the initial and final moments of pouring, cu cm;

F_{sp} = cross section of the sprue, sq cm.

The metal flow rate is

$$Q' = \mu_l F_{sp} \sqrt{2gP} \quad \text{and} \quad Q'' = \mu_l F_{sp} \sqrt{2g(P - L)}. \tag{121}$$

The calculations by the first method are simpler but the cross sections of the channels obtained are larger.

When the flow of metal is not restricted the optimum duration of pouring and the best shape of the stream can be ensured by either method only for one mould. When several moulds are to be filled the head in the ladle and in the gating system should be preferably calculated from the duration of the filling of the first mould, by Eq. (101), with the coefficient S indicated in Table 18, being decreased by 0.2.

The value of S for the last mould should be optimum or 0.1 higher.

The second method. This method finds application when a large number of identical small or large moulds have to be filled with metal, since only one or two first moulds can be filled without shutting off the stream. This method was developed by B. Gulyaev and P. Vasilevsky.

With this method the level of metal is maintained 5 cm below the top level of metal in the pouring cup (level AB in Fig. 86, *3*), i. e., at constant height. The energy of the stream falling from the ladle and striking against the surface of the molten metal is lost almost entirely and affects but little the flow rate of metal in the gating system. This makes it possible to calculate the cross section of the gating system for restricted pouring in the same way as for pouring from tipping ladles by Eqs (106) and (107), i. e., taking into account only the static head and neglecting the energy of the falling stream.

For actual pouring conditions, the size of the nozzle hole in a stopper ladle can be found from Gulyaev's formula

$$F_{sp} = \frac{\pi d^2}{4} \frac{\beta_e}{\beta_r}, \qquad (122)$$

where d = diameter of the nozzle hole;

β_e = erosion coefficient;

β_r = restriction coefficient.

The coefficient β_e accounts for an increase in the nozzle hole cross section during pouring. With the refractories of medium-class quality, the coefficient β_e comes to 1.2-1.7 by the end of the operation. The coefficient β_r accounts for the shutting off of the taphole by the stopper. Observations have shown that at $\beta_r = 1.2$-1.3 the stream retains a good shape, while at 1.6-1.8 spattering and distortion of the stream may occur which will spoil the casting surface. To pour steel effectively into several identical moulds the pouring must be obviously begun at such level of steel in the ladle when the restriction coefficient β_r does not exceed 1.3, while the last mould can be filled at such low level which would ensure the maximum duration of pouring. An excessive increase in the duration of pouring above the maximum value is out of the question.

Let us determine the initial H'_{1min} and the final H''_{1min} heads in the ladle at which the metal can still be maintained at the required height in the gating system.

Transforming Eqs (116) and (117) we shall obtain

$$H'_{1\,min} = h_i \left(\frac{\mu F}{\mu_l F_t} \right)^2 \qquad (123)$$

and

$$H''_{1\,min} = h_f \left(\frac{\mu F}{\mu_l F_t} \right)^2, \qquad (124)$$

where F = cross section of the narrowest spot in the gating system, sq cm;

F_t = cross section of the nozzle taphole, sq cm;

$h_i = h - 5$ cm;

$h_f = m - 5$ cm.

The height of the metal layer expended to fill one mould can be found from the equation

$$\Delta H_1 = \frac{4,000\,G}{\pi D_l^2 \gamma_l}\,, \qquad (125)$$

where G = weight of the metal poured into the mould, kg;

D_l = mean inside diameter of the ladle, cm;

γ_l = specific weight of liquid steel, equal to 7.0 g/cm³.

There is the following relationship between the initial and final levels of steel in the ladle

$$H'_{1\,min} = H''_{1\,min} + \Delta H_1. \qquad (126)$$

To maintain the level of metal in the gating system at the required height during pouring, we must use the larger of the two values of H'_{1min} calculated from Eqs (123) and (126). This will be the value of the minimum initial head in the ladle at which the level of metal can still be maintained at the required height in the gating system and the optimum duration of pouring preserved. According to P. Vasilevsky, we take at $\beta_r = 1.3$

$$H'_{1\,max} = 1.69\,H_{1\,min}, \qquad (127)$$

where H_{1max} is the head in the ladle at which the restriction of the stream at $\beta_r = 1.3$ ensures the same flow rate of steel from the ladle as at H_{1min} and $\beta_r = 1$.

We find:

$$H'_{1\,max} = 1.69\,H'_{1\,min} \qquad (128)$$

and

$$H''_{1\,max} = 1.69\,H''_{1\,min}, \qquad (129)$$

where H'_{1min} and H''_{1min} are the minimum initial and final heads in the ladle found from Eqs (123) and (124).

The final head H'_{1max} determined from Eq. (129) corresponds to the initial head

$$H'_{1\,max} = H''_{1\,max} + \Delta H_1. \qquad (130)$$

In order not to shut off partially the stream of steel falling into the first mould at $\beta_r = 1.3$ during the course of pouring, the smaller of the two values of H'_{1max} found from Eqs (128) and (130) should be used.

The number of moulds poured at the maximum duration and at $\beta_r = 1.3$ will be

$$n = \frac{H'_{1\,max} - H'_{1\,min}}{\Delta H_1} + 1.$$ (131)

Let us use by way of example the calculation of the pouring conditions and of the gating systems for casting steel frames with a total metal flow rate of 1,000 kg per one mould (two castings in one mould). The metal is poured from a stopper ladle 1,800 mm in diameter. The calculation results are represented in Table 18.

Table 18

Number of Moulds Filled from One Ladle at Different Pouring Conditions

(Dubitsky)

Pouring method	Safe values of coefficient S	Rated duration of pouring, sec	Number of moulds filled from one ladle
Unrestricted	1.4-1.6	28-32	6
Restricted	1.5	30	5
Restricted	1.5-1.7	30-34	11

In bottom unrestricted pouring the narrowest section in the gating system can be found from Eqs (110) and (111). The maximum values should be selected.

The minimum weight of the gates is obtained with the most constricted section between F_{sp} and F_g when high quality materials are used for the gating channels. As a rule, $F_{sp}:F_g = 1.3$ to 1.4.

In pressure gating systems intended for steel castings the restricted section is the section of one or several gates F_g or ΣF_g. With account taken of the energy of the falling stream at $F_{sp}:F_g = 1.3$, the area $F_g = F = 33.7$ sq cm and $F_{sp} = 43.8$ sq cm, while at $F_{sp} = F$, the area $F_g = F = 27$ sq cm and $F_{sp} = 27$ sq cm.

Let us consider the condition of an unfilled gating system. In lip pouring (Fig. 86, 4) the metal enters the sprue and the sprue base, gradually fills the runner and streams into the mould cavity. If the ladle is tilted still more the level of metal will rise to ab. The height of the head will then be equal to the difference between the levels ab and MN, i. e., to h', and the linear and weight velocities will be expressed by the equations

$$v_{lin} = \mu 4.43 \sqrt{h' - h}$$ (132)

and

$$v_w = f\mu\, 4.43 \sqrt{h' - h}.$$ (133)

The most important will be the cross section of the sprue ab since the runner is not yet filled. After the mould has been filled to the level $a'b'$ the level of metal in the sprue will begin to rise and the velocity and flow rate will now be determined from Eqs (132) and (133), where h is the metal level in the mould above MN.

The metal will now move in conformity with the principle of flooded opening, i. e., this will be a bottom gated system and the liquid movement will be self-retarding (see Fig. 20).

The head h' is considerably smaller than the head H but should slightly exceed h for, otherwise, the flow of metal will stop. At the end of pouring when the system will turn by itself into a bottom gated system the most important will be the flooded hole, i. e., the section MN and the section of the sprue. Thus, when high moulds are side gated the cross section of the runner should not close up the system and cannot be smaller than the cross section of the sprue. This determines the principle of a subnormal pressure gating system.

Pouring done with the basin kept full, in which the level of metal is determined by the height H (Fig. 86, 5) above section AB will in no way differ during the entire pouring period from the above version of pouring through the cup (Fig. 86, 4), if the section MN is wider than that of the sprue. At the start of pouring the most important will be the section ab of the sprue above which the shape of the falling stream will taper. The sprue will be filled only towards the end of pouring when the runner is completely filled.

The versions 4 and 5 become identical.

At the start of pouring the filling velocities will be determined by Eq. (132) and at the end by Eq. (133). To calculate the cross section of the runner and the sprue we shall use the mean head which tends to bring closer together the different heads at the start and end of pouring.

When the casting is gated through the bottom (Fig. 86,6) with a subnormal pressure gating system, the most important will be the section ab and the head h' and, at the end, the section MN and the head difference $H+h'-h$; accordingly, we shall obtain the following equations

$$v'_{lin} = \mu' \, 4.43 \sqrt{h'}; \quad v'_w = f_{ab} \mu' \, 4.43 \sqrt{h'} \qquad (134)$$

and

$$v''_{lin} = \mu'' \, 4.43 \sqrt{H+h'-h}; \quad V''_w = f_{AB} \mu'' \, 4.43 \sqrt{H+h'-h}. \quad (135)$$

In a pressure gating system each subsequent section grows progressively smaller than the preceding one

$$F_{sp} > F_r > \Sigma F_{g'}. \qquad (136)$$

where F_{sp} = cross section of the sprue, sq cm;

F_r = cross section of the runner which connects the sprue with several gates, in a general case, sq cm;

ΣF_g = sum of cross sections of all gates, sq cm.

The velocity and the weight flow rate are determined by the most constricted spot irrespective of whether it is a pressure or subnormal pressure system.

Dietert's equation will be very helpful for the calculation of the mean head in Eg. (108) when applied for the constricted section

$$H_m = H_{max} - \frac{H_g^2}{2H_{cast}} = \frac{2H_{max}\,H_{cast} - H_g^2}{2H_{cast}}, \qquad (137)$$

where H_m = mean pressure head, cm;

H_{max} = maximum head (by the level of metal in the pouring basin above the gate), cm;

H_g = height of casting above the gate, cm;

H_{cast} = total height of the casting with the mould arranged for filling, cm.

With top gating (Fig. 86, 1) $H_g=0$ and

$$H_m = H_{max}. \qquad (138)$$

With the metal poured from the basin (Fig. 86, 2)

$$H_g < H_{cast} \text{ and}$$

$$H_m = H_{max} - \frac{H_g^2}{2H_{cast}}. \qquad (139)$$

With bottom gating (Fig. 86, 3, 4, 5 and 6) $H_g = H_{cast}$ and

$$H_m = H_{max} - \frac{H_{cast}}{2}. \qquad (140)$$

To establish the pouring rate for the moulds fed with metal from a stopper ladle, through the bottom, the nozzle diameter should be selected first and, then, the weight flow rate of steel from the ladle equated to the mould pouring rate by the group principle. Small castings are poured from small ladles and large castings through two and more sprues. Very small castings in large-lot production are poured from a tea-pot spout ladle.

The dimensions of stopper ladles are illustrated in Table 19.

The flow rate coefficient or, in other words, the coefficient of hydraulic resistance is

$$\mu = \frac{1}{\sqrt{1 + \Sigma \xi_i}}, \qquad (141)$$

where $\Sigma \xi_i$ is the sum of hydraulic resistances in the way of liquid flow.

Table 19

Dimensions of Stopper Ladles

Weight of charge, tons	Volume of metal and slag (10%), cu m	Inside diameter, m		Total height, m
		at top	at bottom	
10	1.81	1.69	1.60	1.69
15	2.73	1.84	1.75	1.84
20	3.64	2.13	2.02	2.13
25	4.55	2.30	2.18	2.30
30	5.46	2.44	2.32	2.44
35	6.37	2.54	2.44	2.57
40	7.28	2.68	2.54	2.68
45	8.19	2.80	2.66	2.80
50	9.10	2.88	2.74	2.88

S. Birulya determined the resistance for a 90° turn of the channel for pig iron in a sand mould and found that $\xi = 1.9\text{-}2.5$. At a turn of 45°, $\xi = 0.6\text{-}0.8$ and at 135° it was about 3.8-5.0. When the channel was branched off at an angle of 45° to the direction of flow, the resistance coefficient amounted to about 120 per cent, at an angle of 90° it was 150 per cent and at an angle of 135°—about 300 per cent, relative to the losses incurred in case of the 90° turn.

The coefficient of resistance in case of a sudden widening of the channel can be found from the formula

$$\xi = \left(1 - \frac{f_1}{f_2} \right)^2, \tag{142}$$

where f_1 and f_2 are the cross sections of the channel before and after widening.

When the ingate feeds the metal into the mould cavity, f_1 is extremely small as compared to f_2 and, for this reason, the value of the second term in the brackets may be assumed equal to zero, and $\xi = 1$. The total losses due to the variation in the cross sections of the gating system amount, as a rule, to 0.2-0.4.

The friction losses in the channels can also be determined with the help of hydraulics formula

$$\xi = \lambda \frac{l}{d}, \tag{143}$$

where λ = dimensionless factor which is a function of the Reynolds number and the degree of the channel roughness;

l = channel length, cm;

d = channel diameter, cm.

The value of the coefficient μ for steel castings has been studied by various investigators. The data they obtained are represented in Table 20.

Table 20

Coefficient μ for Steel Castings

Casting system	Coefficient μ	Source
Medium	0.35-0.48	Bidulya
Complex	0.25-0.30	Nekhendzi
Medium	0.30-0.40	Nekhendzi
Simple	0.40-0.50	Nekhendzi
Complex	0.25	Dubitsky
Medium	0.32	Dubitsky
Simple	0.42	Dubitsky
Medium	0.80-0.41	Osann

The experience gained by foundrymen allows certain rules to be recommended in designing gating systems for steel castings.

1. The metal should be fed to the mould and across its cavity by the shortest routes possible to reduce the metal consumption and prevent its solidification in the stream and in the faraway sections of the casting, to lessen the hazard of erosion of the channel and the mould walls.

2. The ingates should never be cut in near the protruding parts of the mould to avoid the metal stream striking against the vertical wall of the mould. The stream should be directed along the longitudinal axis of the mould wall.

3. The design of the gating system channels must provide for narrowing of the stream, i. e., the most constricted section should be in the ingates and not in the sprue to raise the pressure inside the stream above the atmospheric. Otherwise, the stream will suck in the air and gases thereby promoting the formation of gas holes in the casting.

4. The principle of consecutive filling of a complex mould should be strictly adhered to: the first portions of metal should be bottom gated, the next ones delivered under the flooded layer and the upper risers filled at the end through the upper level ingates cut in under the risers (see Fig. 48).

5. The design of the gating system should provide for a good directional solidification of the casting. In other words, local

thickened sections should not be heated by special ingates but cooled with local chills. The risers should be filled with hot metal fed over the shortest routes or else heated by thermit core inserts. The ingates should never be allowed to feed the bottom of the whole casting or feed the risers in case of large-size castings at the end of pouring, during contraction and solidification.

6. The ingates should never be arranged near the core prints, as shown in Fig. 87. Ingate *2* brought close to core *1* on the diagrams *a, b* or *c* will produce fins in gap *3* between the core and the
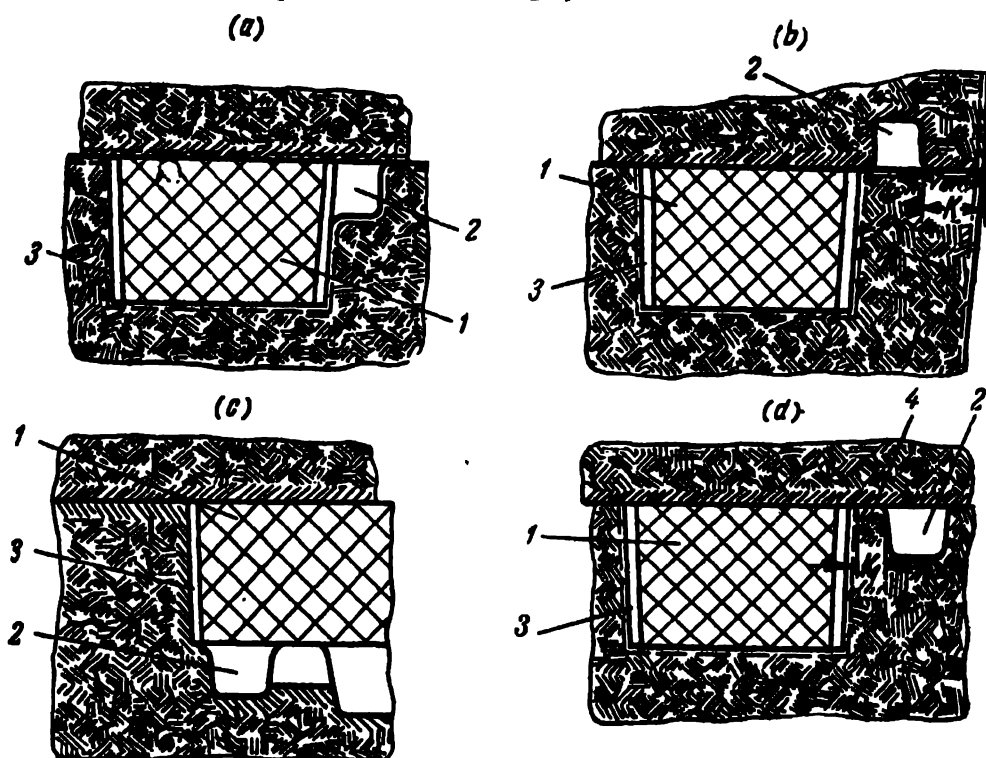
Fig 87. Location of ingates near the core in the mould (Vasilevsky):
1—core; *2*—ingate; *3*—gap; *4*—wall

mould. The fins complicate the finishing operation and may cause local cold cracking.

In the diagram *c* the core may crumble while being inserted into the mould and clog up channel *2* with dirt. In the diagram *d*, sand wall *4* may disintegrate because it is too weak and cause similar defects. The safe minimum distances between the gating channels and the adjacent parts of the mould as illustrated in Fig. 87 are indicated in Table 21.

7. The ingates should never be arranged at the chills or studs as this may quickly dissolve the studs in molten steel and minimise the anticipated effect.

8. Attempt should be made to ensure a unidirectional movement of the molten metal in the mould and a rotary movement in the bodies of revolution in horizontal planes of the castings to

Table 21

Minimum Safe Distances Between the Gating Channels and Adjacent Parts of the Mould, mm

(Vasilevsky)

Size, mm	Mould	Weight of casting, kg											
		up to 50	50-250	250-500	over 500	up to 50	50-250	250-500	over 500	up to 50	50-250	250-500	over 500
		Gate length				Distance K_1				Distance K			
up to 25	Dry	40	50	—	—	30	35	—	—	30	35	—	—
	Green	55	60	—	—	50	60	—	—	30	35	—	—
25-50	Dry	45	55	60	70	40	40	45	50	40	40	45	50
	Green	60	65	70	80	50	55	60	70	45	50	55	60
Over 50	Dry	55	60	70	80	50	50	55	60	50	50	55	60
	Green	60	70	80	90	60	65	70	80	60	60	65	70

facilitate the upflow of the nonmetallic inclusions and films to the risers or padding provided in the upper parts of the mould.

9. Metal should be fed to thin-walled and large-size castings liable to develop cold cracking as far as possible through several gates and the thick sections be cooled with local chills as, for example, when casting gear wheels or sheaves.

10. The cross section of the gating channels should have the minimum, annular, cooling surface, while the gates for thin-walled castings may be of a slot-type design.

11. Gating channels should be provided in refractories only in case of very large castings and when they are absolutely indispensable for obtaining sound castings.

19. RISERS. DESIGN, SHAPE AND SIZE. CALCULATION METHODS

The process of cooling and solidification of molten alloys is attended by volumetric contraction which results in the appearance of shrinkage cavities. The function of open and blind risers (Fig. 88) is precisely to compensate for this shrinkage by feeding metal to the casting, thereby ensuring its solid structure. After shakeout the riser is cut off from the casting and remelted. Hence the important task of reducing the amount of metal that goes to make the riser to the maximum so as to increase the ratio between the weight of good ingot and the combined weight of the metal being cast and the riser. When the riser is too small it will fail to

feed the casting adequately. Then, not only the riser but the defective casting too will have to be remelted. If the riser is excessively large extra metal will have to be spent to make a sound and solid casting. It follows from this that the designer should select such size and shape of the riser as would produce a solid casting free from shrinkage cavities and porosity at the minimum weight of the riser and with the most economical metal consumption.

It should be borne in mind that the riser can serve its purpose only in conditions of .good directional solidification. In calculating the shrinkage porosity, according to I. Kumanin and B. Gu-
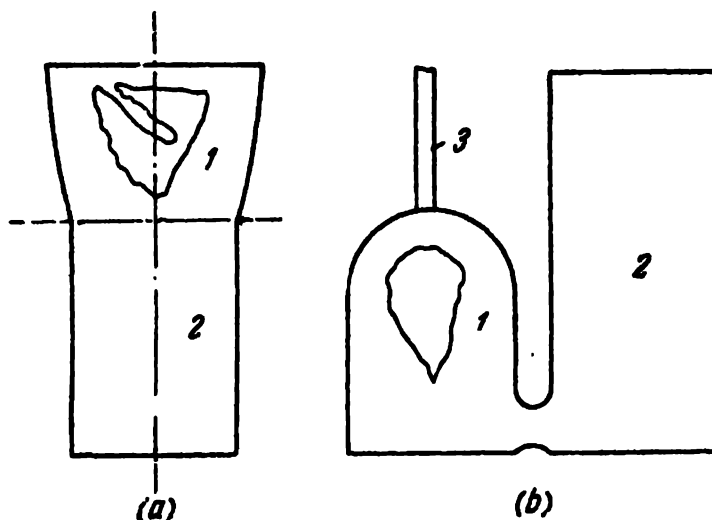
Fig. 88. Riser shapes:
a—open cylindrical; *b*—blind semispherical; *1*—riser; *2*—casting; *3*—pop-off went

lyaev, it was pointed out that no risering can eliminate completely shrinkage porosity which depends on the temperature range of solidification of the alloy and the heat accumulating ability of the mould. This has been confirmed by experiments conducted by Pellini, which show that shrinkage porosity in a plate and a bar of equal cross section in length will not diminish even if the riser volume is increased sevenfold.

V. Grum-Grzhimailo was the first to put forward this maxim in designing risers for steel castings: "The riser must solidify last".

B. Gulyaev believes that a riser must have:

(1) sufficient cross section to be able to solidify after the casting;

(2) sufficient volume so that the shrinkage cavity may not break through to the surface;

(3) a shape ensuring the minimum volume.

The first two requirements are analogous to the maxim of V. Grum-Grzhimailo, as the shrinkage cavity will not break through the riser only if it solidifies last.

Foundrymen use both open and blind risers (Fig 88). The latter type is more expedient because less heat is lost at the top and less dirt penetrates into the metal, the area of the open channels of the mould being smaller. Ordinarily, blind risers are provided with pop-off vents to permit the escape of gases from the mould during pouring.

Heat transfer arises in the mould as the alloy is being cooled, the mould absorbing the heat given off by the alloy. The more metal there is per unit of the heat absorbing surface of the mould, the longer is the process of solidification. A spherical or semispherical riser is, therefore, the most rational design since, with an equal volume, they have the minimum cooling surface and, hence, the maximum referred thickness as compared with differently shaped risers.

Spherical risers are used with great advantage for the production of steel anvil blocks. However, the complicated moulding they involve somewhat limits the sphere of their application.

A semispherical riser which is gaining an increasingly broader application today is the most expedient design. It carries a pop-off vent on the top to permit the gases to escape and the steel to fill the riser completely.

The maxim of V. Grum-Grzhimailo can be expressed by the following inequality

$$\tau_{sol.\ r} > \tau_{sol.\ c}$$

or with account taken of Eq. (45)

$$M_r \Phi_r R_r^2 > M_c \Phi_c R_c^2. \qquad (144)$$

If the riser and casting are of the same material, $M_r = M_c$ and Eq. (144) assumes a simpler form

$$\Phi_r R_r^2 > \Phi_c R_c^2 \qquad (145)$$

or

$$R_r > \sqrt{\frac{\Phi_c}{\Phi_r}}. \qquad (146)$$

For the plate $\Phi = 1.00$, for the cylinder and bar $\Phi = 0.76$, for the ball and cube $\Phi = 0.47$.

With a spherical riser ($\Phi = 0.47$), at a given value of R_r and with a casting in the form of a plate ($\Phi = 1.00$), the right side in the inequality (146) will diminish only in case of R decreasing. This can be attained by two means: by increasing the plate thickness and reducing its sides, or decreasing its thickness and increasing the length of the sides. In the first case the plate volume will be smaller and in the second greater. Under these conditions the volume of the riser will remain constant. Hence the following important conclusion: with the same referred thickness of the casting an increase in its cooling surface (the second case of R_r

decreasing) will promote the efficiency of the riser, since this increases the yield of good ingot (a riser of the same weight feeds a heavier casting). This accounts for the fact that with a broad cooling surface of the casting the weight of the riser amounts to 20-25 per cent of the casting weight, whereas for a casting approximating in shape to a cube or ball its weight will be 100 per cent and more of the weight of the casting.

To feed the casting with liquid metal correctly the designer should first determine the dimensions of the risers (to conform to their geometry) and their total number.

Today various methods are available for an approximate calculation of risers, which can be subdivided into two main groups:

(1) methods based on direct and indirect determination of the solidification time of the riser and casting;

(2) methods based on determining the height of the shrinkage cavity in the riser.

We shall confine ourselves to the simplest methods verified under different production conditions.

Briggs believes that, depending on calculated solidification time of the casting and riser, the diameter of the latter should amount to $1.1T$ for the cube, $1.6T$ for the bar and $2.5T$ for the plate (where T is the casting thickness). The height-to-diameter ratio of an open riser should not be below 1.5. In this case all the risers should be heated to obtain the necessary temperature difference and good progressive solidification.

A. Popov bases his calculations on the square root equation and, although there is a certain error in the calculation of cylindrical risers, it does not affect essentially the determination of their size.

The method is as follows. According to N. Chvorinov, the mean thickness of the skin solidified during the pouring time τ_1 is

$$\xi_m = \frac{k \sqrt{\tau_1}}{2}. \qquad (147)$$

If the casting surface is S, the volume of the metal during the pouring process will be:

solidified metal

$$V_{sol} = \xi_m S, \qquad (148)$$

liquid metal

$$V_l = V_0 - V_{sol}. \qquad (149)$$

Assuming the volumetric shrinkage of steel in solidification to be equal approximately to 3 per cent we shall obtain the volume of shrinkage cavities in the casting

$$V_{c.c} = 0.03 (V_0 - V_{sol}). \qquad (150)$$

If we disregard the outflow of metal from a cylindrical riser to feed the casting and neglect the amount] of metal solidifying in it during pouring, the volume of the shrinkage cavity can be found from the equation

$$V_{c.r} = 0.03V_r = 0.03 \frac{\pi d^2}{4} h = 0.075 \pi d^2 h, \qquad (151)$$

where d = diameter of the riser base determined by the method of inscribed circles or by other methods and increased by about 20-30 per cent as compared with the diameter of the part of the casting fed by the riser;
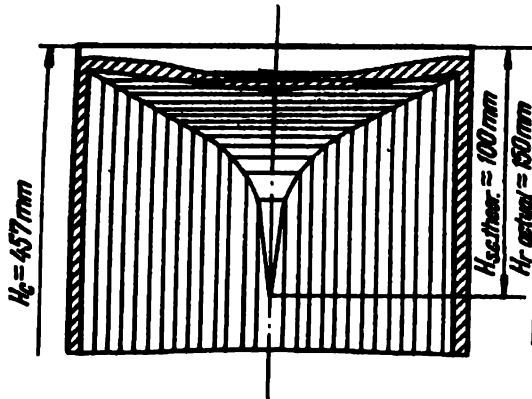
Fig. 89. Graphical scheme illustrating the formation of a shrinkage cavity in a cylindrical riser intended for a casting 150 mm in diameter and 457 mm long (Popov) .

h = unknown height of the shrinkage cavity in the riser.

The total volume of the shrinkage cavity in the riser will be comprised of the volumes of the shrinkage cavities in the casting and the riser:

$$\Sigma V_{s.c} = V_{c.c} + V_{c.r} = 0.03 (V_0 - V_{sol}) + 0.0075 \pi d^2 h. \qquad (152)$$

The shrinkage cavity develops in the riser in conformity with the law of parabola, as was shown graphically by A. Popov (Fig. 89). The form of the shrinkage cavity as a spheroidal cone is distorted by the open riser being cooled from the top and more intensive cooling of its upper corners. Therefore, the isotherms are curved as shown schematically in Fig. 90 and the form of the cavity approaches a right cone with the base 0.5 d at bottom gating and 0.6 d at top gating. The cone base in a spherical riser is 0.5 d.

Regarding the actual form of the shrinkage cavity as a right cone and expressing its volume in terms of the riser diameter, Eq. (152) can be written as follows:

$$\Sigma V_{s.c} = \frac{\pi (nd)^2 h}{12} = 0.03 (V_0 - V_{sol}) + 0.0075 \pi d^2 h, \qquad (153)$$

where n = ratio between the cone base and the riser diameter.

Formulas (147) and (148) can be used to solve this equation with respect to h

$$h = \frac{0.0575 \, (2V_0 - kS \, \sqrt{\tau_1})}{(n^2 - 0.09) \, d^2} . \qquad (154)$$

In case of short-time pouring of cube- or ball-shaped castings with a high volume-to-surface ratio, the expression $kS\sqrt{\tau_1}$ tends
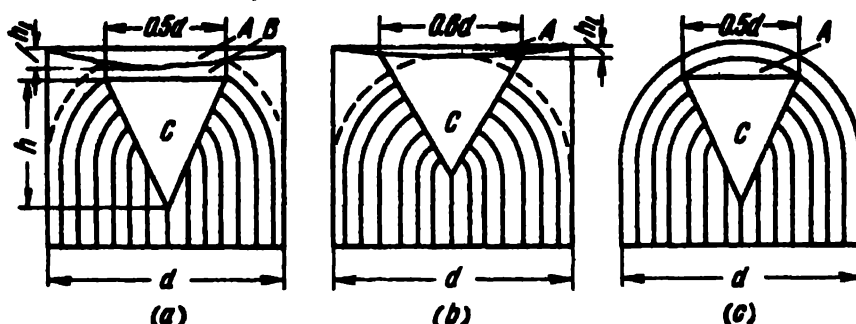
Fig. 90. Shape and location of shrinkage cavities at the surface and below the surface:

a and *b*—cylindrical riser; *c*—semispherical riser (Popov); *A*—shrinkage cavity above the riser; *B*—bridging over the shrinkage cavity; *C*—shrinkage cavity below the surface

to zero and can be neglected. This makes it possible to simplify the calculations and impart to Eq. (154) a form common for sand-mould and metal-mould castings

$$h = \frac{0.115 \, V_0}{(n^4 - 0.09) \, d^2} . \qquad (155)$$

In large-size castings of small volume but with a large surface the term $kS\sqrt{\tau_1}$ is very near to $2V_0$ and the expression in the brackets in the numerator of Eq. (154) becomes zero. The same occurs at slow rates of pouring. In this case the risers are no longer needed.

The actual height of the riser for steel castings is (Fig. 91)

$$H_r = h + h_{s.c} + h_{br} + h_{con}, \qquad (156)$$

where H_r = total height of the riser, cm;

h = height of the shrinkage cavity in the riser, cm;

$h_{s.c}$ = height of the spherical concavity (shrinkage cavity at the surface), cm;

h_{br} = thickness of the bridging (solid metal) over the cavity, cm;

h_{con} = reserve height of the riser for loose metal under the tail of the shrinkage cavity and for contamination by rising nonmetallic inclusions, cm.

As a rule, $h_{s.c} + h_{br} + h_{con} = 4$ cm.

A. Popov assumes the solidification constant to be equal to 0.645 cm min $^{1/2}$ for a sand mould and 2 cm min $^{1/2}$ for a metal mould.
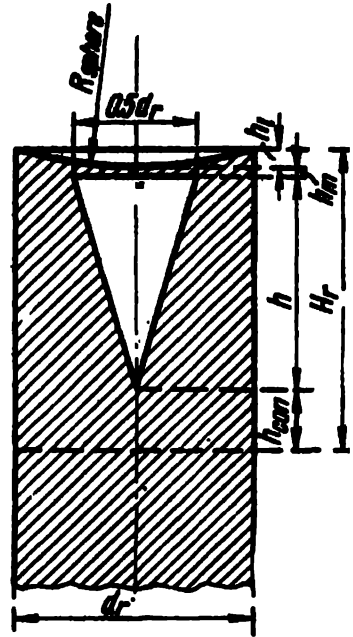
Fig. 91. Calculating the riser height (Popov)

Proceeding from the above and using Eqs (154) and (155), the total height of the riser for castings with volume V_0 and surface S is finally determined from the following equations:

(a) bottom gated sand moulds

$$H_r = \frac{0.23 \, (3V_0 - S \, \sqrt{\tau_1})}{d^2} + 4 \text{ cm};\qquad(157)$$

(b) heavy castings bottom gated (by simplified equation) into sand moulds

$$H_r = \frac{0.7 \, V_0}{d^2} + 4 \text{ cm};\qquad(158)$$

(c) top gated sand moulds

$$H_r = \frac{0.15 \, (3V_0 - S \, \sqrt{\tau_1})}{d^2} + 4 \text{ cm};\qquad(159)$$

(d) heavy castings top gated into sand moulds

$$H_r = \frac{0.45 \, V_0}{d^2} + 4 \text{ cm};\qquad(160)$$

(e) bottom gated metal moulds

$$H_r = \frac{0.7 \, (V_0 - S \, \sqrt{\tau_1})}{d^2} + 4 \text{ cm};\qquad(161)$$

(f) heavy castings bottom gated into metal moulds

$$H_r = \frac{0.7 \, V_0}{d^2} + 4 \text{ cm};\qquad(162)$$

(g) top gated metal moulds

$$H_r = \frac{0.45\,(V_0 - S\,\sqrt{\tau_1})}{d^2} + 4 \text{ cm};$$ (163)

(h) heavy castings top gated into metal moulds

$$H_r = \frac{0.45\,V_0}{d^2} + 4 \text{ cm};$$ (164)

(i) bottom gated metal moulds with a heated ceramic runner bush

$$H_r = \frac{0.3\,(V_0 - S\,\sqrt{\tau_1})}{d^2} + 4 \text{ cm};$$ (165)

(j) top gated metal moulds with a heated runner bush

$$H_r = \frac{0.2\,(V_0 - S\,\sqrt{\tau_1})}{d^2} + 4 \text{ cm}.$$ (166)

The last two equations are simplified.

American researchers suggested a graphical method of calculating the risers (Fig. 92) heated with exothermic sleeves. They believe that the riser-to-casting volume ratio is determined by the shape coefficient $\frac{L+W}{T}$ where L is the length, W—width and
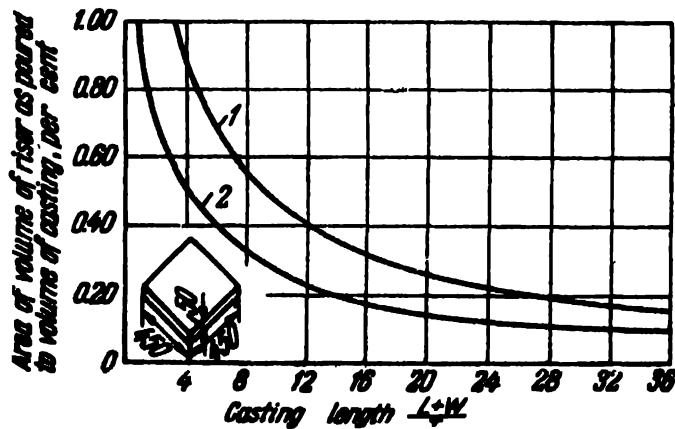
Fig. 92. Change in the riser-to-casting volume ratio as a function of the casting length $\frac{L+W}{T}$

(Bishop, Johnson):

1—upper limit; *2*—lower limit

T—thickness of the casting. The riser is of an open and cylindrical design. The height-to-diameter ratio is 0.5-1. To avoid rejects the volume of risers should be assigned by the upper curve and then, if warranted by production practice, the volume may be decreased but not below the lower curve. The riser height is determined by Fig. 93 (inclined lines—riser diameter) where the upper curve is for the ratio $\frac{H_r}{D_r} = 1$ and the lower curve for $\frac{H_r}{D_r} = 0.5$

When designing a riser for a shaped casting it should be separated into simplest heat centres requiring independent feeding. The shape coefficient is determined from the basic dimensions of the casting and disregards various branch pieces connected to the bulk of the casting. The riser is positioned over the heaviest
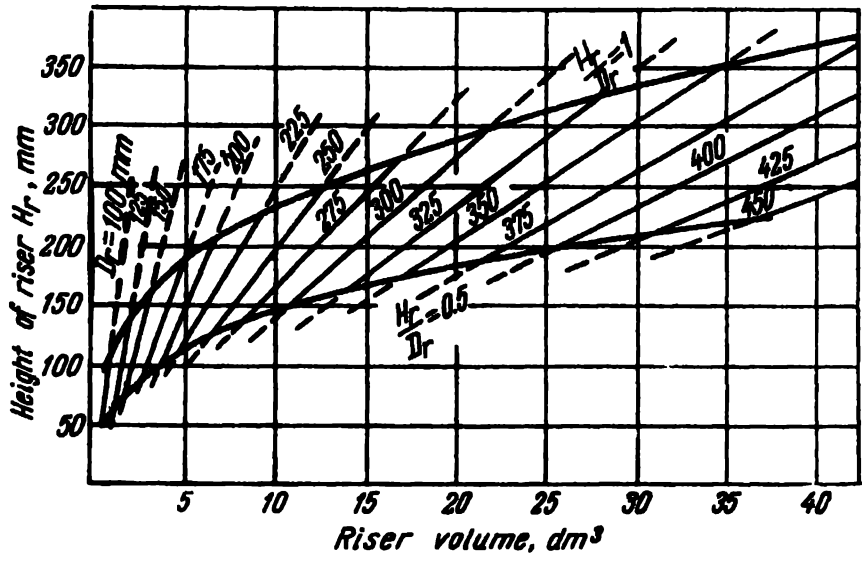
Fig. 93. Change in the riser height as a function of its volume (Bishop, Johnson)

section and, if a branch piece is attached to it, the riser should be increased in volume to feed both. For this purpose, a special correction depending on the shape of the main section and the branch piece found as shown in Fig. 94 should be introduced. For example,
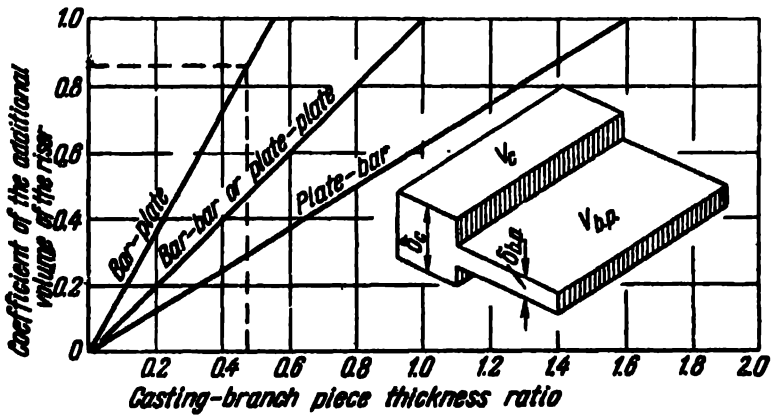
Fig. 94. Graph for determining additional volume of the riser to feed branch piece (Bishop, Johnson)

in a bar-shaped casting 100 mm thick with a plate-shaped branch piece 50 mm thick attached to it, the ratio $T_{b.p}$: T_c is 0.5. By erecting a vertical line to the inclined line (bar-plate) and drawing a horizontal line to the Y-axis we shall obtain the coefficient of

the additional volume of the riser. It is 0.85. The total volume of the casting $V_c = V_{b,p} + 0.85 \, V_c$. The shape coefficient will be found from the bar size. The size of the riser can be found from the data in Fig. 92.

When a casting of uniform cross section is very long the shrinkage porosity can be prevented by arranging several risers. The knowledge of the effective feeding distance of the riser will be extremely
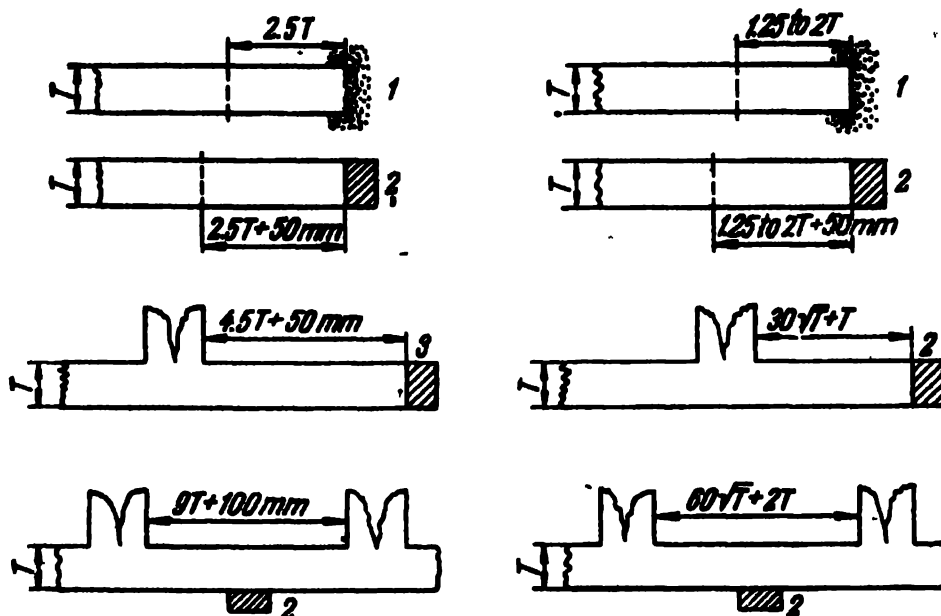
Fig. 95. Effective feeding distance of risers in plates and bars with chills placed at the centre or on one end (Bishop, Myskowski, Pellini):

1—sand; *2*—chill

helpful in reducing the amount of extra metal. Investigations have shown that the feeding distance depends on the shape and thickness of the casting. The distance decreases in a thinner casting. In a bar the feeding distance is smaller than in a plate. The feeding distance of the riser can be increased by the use of chills (Fig. 95). The effective feeding distance was established by gamma-raying of the casting at a sensitivity of 1.5 per cent. If the sensitivity amounts to 4-5 per cent the feeding distance can be increased twofold as against the data in Fig. 95

Let us use this method to calculate the risers for certain typical castings.

Annular castings, such as gears, are very frequent in foundry practice. But the risers used for them require extra metal. Conversely, if the volume or number of the risers is insufficient the casting may develop shrinkage porosity. The volume of a gear rim with a circumference 3,900 mm long and a cross section of 300 × ×100 mm (Fig. 96) is 117 cu dm. With one riser, the volume of the casting being fed also equals 117 cu dm, and the shape coef-

ficient is $\frac{3,900+300}{100} = 42$. After extrapolating the data in Fig. 92, the relative volume of the riser will be 0.12 or $117 \times 0.12 = 14$ cu dm. With this volume and a diameter of 300 mm, the riser height will come to 200 mm (Fig. 93).

If six risers are placed at intervals along the rim the volume of the section being fed will be 19.5 cu dm, with the shape coefficient of $\frac{3,900:6+300}{100} = 9.5$. According to Fig. 92, the volume of one riser is $19.5 \times 0.42 = 8.2$ cu dm, and the volume of all risers
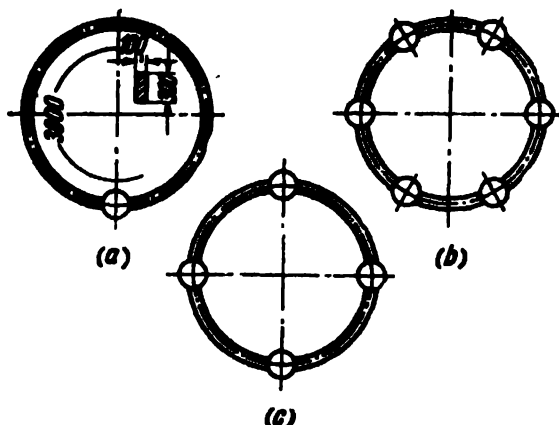
Fig. 96. Positioning of risers along a steel rim (Bishop, Myskowski, Pellini):

a—one riser 300×200 in dia; *b*—four risers 275×210 in dia; *c*—six risers 250×175 in dia

will be 49.2 cu dm. With a diameter of 250 mm (Fig. 93) the riser height will be 175 mm.

When the chills are placed at the centre of the casting the number of the risers can be reduced to three. Then the volume of the section being fed will be 39 cu dm, with the shape coefficient 16. The volume of one riser is $39 \times 0.27 = 10.6$ cu dm and that of all risers 31.8 cu dm. The riser is 275 mm in diameter and 175 mm high.

In the first case a large part of the casting will develop porosity because the length of the part to be fed $\left(\frac{3,900-300}{2} = \right.$ $= 1,800$ mm) will be considerably in excess of the effective feeding distance of the riser which, as shown in Fig. 95, is 400 mm (the rim is 100 mm thick). With six risers, the length of the section to be fed is the same as the riser feeding distance ($400 \times$ $\times 6 + 250 \times 6 = 3,900$ mm) and the ingot is quite good although extra metal is expended on risers.

The design with chills is most rational. The riser feeding distance ($9 \times 100 + 100 = 1,000$ mm) is practically the same as the length of the casting being fed $\left(\frac{3,900-275 \times 3}{3} = 1,025 \right.$ mm), resulting in a good ingot and a minimum expenditure of metal on the risers.

A. Simkin, A. Gulin, V. Zlodeyev and the author experimentally verified the work of the American researchers on thinner

plates and bars (5, 10, 15 and 20 mm). Solid plates 5 mm thick were obtained without risers. In plates and bars 5 and 10 mm thick (see Fig. 95) the feeding distance was $10T$ with chills placed on one end.

Sleeves are very common articles of foundry production (Fig.97). In this case, when the risers are positioned on the flanges, calculations are done only for one half of the casting. The flange with the riser is regarded as the main part in the form of a bar 63 mm thick. For the branch piece is taken a half-cylinder regarded as a plate 19 mm thick. The volume of the flange is 3.85 cu dm, its mean circumference 669 mm; the volume of the half-cylinder is 1 cu dm. The coefficient of the flange shape is $\frac{669+63}{63} = 12$.
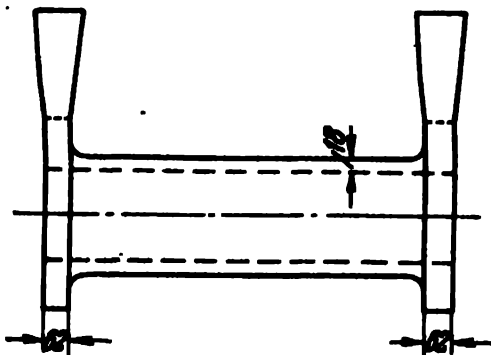
Fig. 97. A steel sleeve with open risers on its flanges (Bidulya)

The ratio $\frac{T_{b \cdot p}}{T_c} = \frac{18}{63} = 0.3$. On the basis of the data in Fig. 93, the additional volume of the casting will be 0.5 cu dm; the volume of one riser will be, as shown in Fig. 92, 0.4 (385+0.5)=1.74 cu dm, and the volume of all risers 3.48 cu dm.

When casting sleeves a correction should be introduced to account for the thickness of their walls (Table 22) since the heat accumulated by the central core delays the solidification of the casting.

Table 22

Correction Factors for Wall Thickness of Sleeves

(Bishop, Myskowski and Pellini)

Core diameter	Correction factor	Rated wall thickness
$0.5T$	1.17	$1.17T$
$1T$	1.14	$1.14T$
$2T$	1.10	$1.10T$
$4T$	1.02	$1.02T$

The data collected by W. Trommer on designing ordinary and heated risers also deserve mention (Figs 98 and 99). In both cases the height of open risers is 1.5 of the diameter.
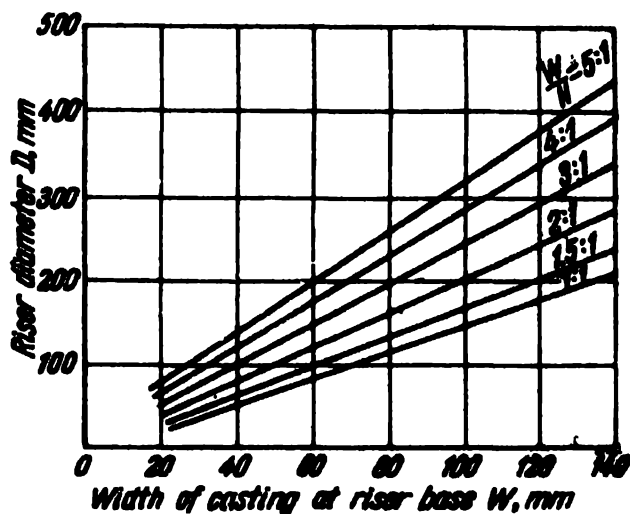
Fig. 98. Change in the riser diameter as a function of its base width (Trommer)

R. Namur developed one of the most precise, although some-what cumbersome, methods for calculating risers. The riser volume is determined from the equation

$$V_r = AR_c^3 + 3\alpha V_c, \qquad (167)$$

where $A = \dfrac{V_c}{R_r^3}$ is the coefficient depending on the shape of the riser;

R_r = referred thickness of the riser;

R_c = referred thickness of the casting or its heavy section connected with the riser;

α = coefficient of steel volumetric shrinkage;

V_c = volume of the casting being fed by the riser.
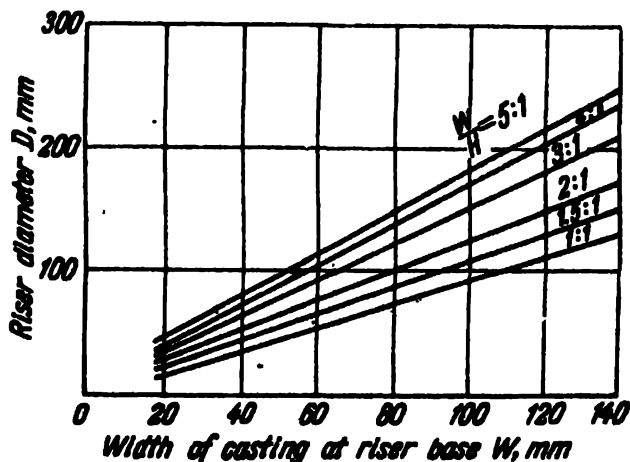
Fig. 99. Change in the diameter of a heated riser as a function of its base width (Trommer)

For a semispherical riser $A = 280$. For a low-carbon steel $\alpha = 0.06$.

All these methods for riser calculation are clearly approximate and largely empirical.

So far, there is no theory of feeding steel castings with risers. This can be explained by the absense of accurate thermotechnical calculations. The approximate calculations utilised by various investigators in certain specific cases were not corroborated by experimental data or produced no reliable results when the shape of the riser or the metal feed conditions were changed.

Soviet scientists have verified the calculations used by Caine, Přibyl and others for riser designing. The results of this verification have shown that, depending on the calculation method, the volume of the riser may vary within broad limits, providing or failing to provide adequate feeding. Experiments conducted with different risers have confirmed the fact that spherical or semi-spherical designs are the best (Fig. 88), while the horizontal cylindrical design shown in Fig. 100 has been found most irrational.
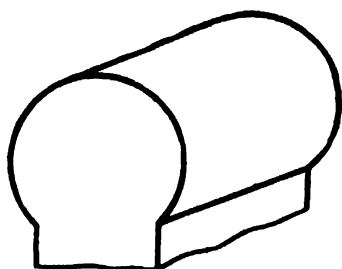
Fig. 100. Horizontal cylindrical riser is the most irrational design (Saveiko)

In this way, the above calculations can be used to arrive at a correct solution, which will meet practical requirements, only in conjunction with gamma raying of test castings. Other methods of control are less reliable but still can be used for uncritical castings.

20. HEATING THE RISERS

As was frequently pointed out before, the positive temperature gradient between the riser and the casting is the primary condition for adequate feed of metal from the riser to the casting during contraction. The feed of metal through long runners (the metal cools somewhat during passage) and the supply of metal to the riser over the shortest routes are the simplest methods used to heat the risers. Large castings are frequently poured with an interruption: at first the entire mould is filled through the sprue to the bottom of the risers and then the risers are directly filled from the ladle.

The moulds for anvil blocks, rolls and other heavy castings which require many hours to solidify should be refilled. In this case, some time is allowed to elapse until volumetric shrinkage becomes quite noticeable. After that the upper solid skin in an open riser is pierced and some more metal is added.

To illustrate. A die bed (0.8×1.2×2.4 m) is cast in a sand mould. The time of complete solidification is $0.592 \times 1 \times 40^2 =$ $= 948$ min or 15.8 hr. The riser for this casting was refilled three

times: for the first time 6 hours after filling the mould, for the second 9 hours later and for the third 12 hours later. The last portion of metal could be poured only after the skin was cut through with a large acetylene torch. During the first repouring the mould consumed about 10 tons of liquid metal, during the second about 3 tons and during the last about 1.5 tons.

In addition to this method, it is frequent practice at large engineering plants to use for large and medium castings certain exóthermic compounds applied to the surface of molten metal in the riser, such as dry sand mixed with charcoal, coke smalls and other combustibles.

Y. Nekhendzi mentions the suggestion of the well-known inventor of electric welding N. Slavyanov, made in 1898, to use an electric arc for heating the risers for large castings and ingots. This method has found a rather widespread application today.

However, when open risers are heated with a voltaic arc or by adding lunkerite or other exothermic compounds, such as aluminothermic, for instance, so much heat is lost that neither the cost of saved steel nor the advantages accruing from less spoilage can fully compensate for it. The best results are obtained from the use of inert insulating sleeves and exothermal sleeves with compounds that evolve the necessary amount of heat. The Foseco Feedex mixture has found a wide recognition in England for this purpose.

These sleeves are available in a great variety of types. The principle underlying their design and size calculation by the amount of heat evolved by the thermal mixture is illustrated below.

Upon combustion the mixture should form readily floating porous slag capable of thermally insulating the riser on the top, without affecting the chemical composition of steel. It must ensure a slow and uniform course of the exothermic reaction, possess good mouldability, produce no burns on the riser surfaces, in no way hamper their removal and, finally, be quite safe in operation.

Numerous experiments helped to find such mixture. It consists of aluminothermic and grade 400 portland cement with no more than 2.5 per cent SO_3 and with a high CaO content.

The highest efficiency was observed in a mix composed of 3 parts (by weight) of aluminothermic and 1 part of grade 400 portland cement with 10-15 per cent of water (per dry mass). The more moisture there is in the mix, the more slowly it heats up and better preserves its plasticity and mouldability. Setting occurs 1 to 1.5 hr after mixing with water. The sleeves need not necessarily be dried. The samples held for twenty four hours, to which some more water was added afterwards, exhibited a somewhat reduced strength and moulding plasticity. The specific weight of a compacted mix was 1.9-2 kg/dm³ and the specific heat of combus-

tion—650 kcal/kg. The combustion of a unit weight of the mixture thus composed produces 60 per cent of porous slag and 40 per cent of metal containing 75 per cent of aluminium and 25 per cent of iron ore.

The experiments have shown that the efficiency of the mix amounts to about 0.40 of the obtained heat on combustion in carbon and to 0.45 in alloy steel. When a thermal mix without inhibitor and binder was applied the efficiency was only 0.05, while the rest of the heat was dissipated by radiation.
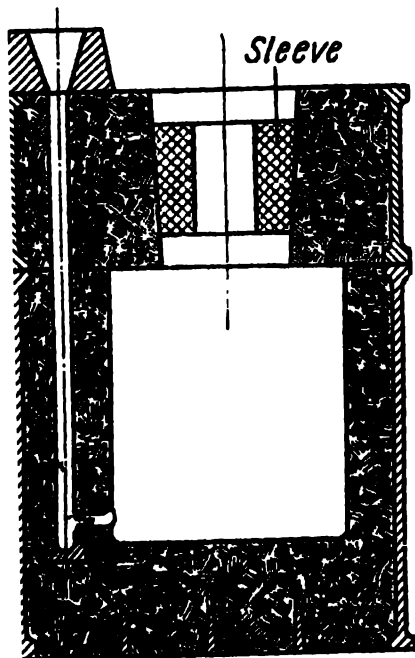
Fig *101.* Riser with a thermit sleeve (Bidulya, Manakin)
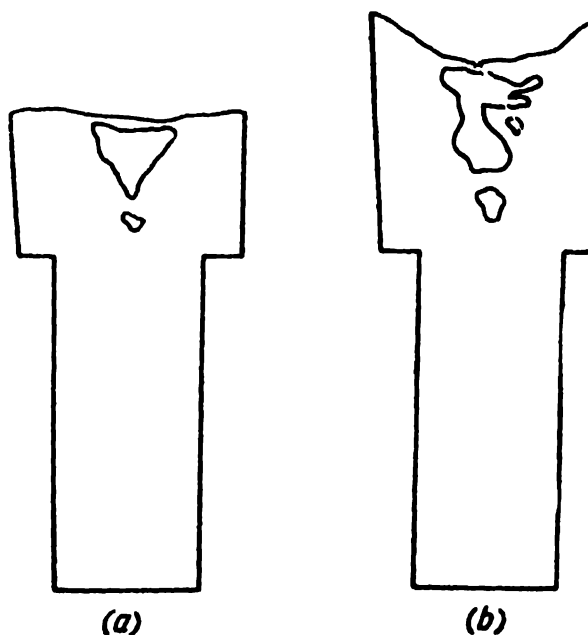
Fig. *102.* Castings (cross section) of heat resistant steel (Bidulya, Manakin):
a—with a heated riser; b—with an ordinary rise

The location of a sleeve in the riser for a steel casting is shown in Fig. 101. Castings with an ordinary riser (right) and a heated riser (left) are illustrated in Fig. 102.

The heat content of an ordinary riser is determined by the heat capacity of the molten metal and the latent heat of solidification

$$Q_{or} = G_{or} [c' (\theta_p - \theta_s) + \varrho] \text{ kcal,} \qquad (168)$$

where Q_{or} = heat content of an ordinary riser, kcal;
G_{or} = weight of an ordinary riser, kg;
c' = heat capacity of molten metal, kcal/kg per deg;
θ_p = temperature of molten metal during pouring, °C;
θ_s = solidus temperature, °C;
ϱ = latent heat of solidification, kcal/kg.

When the weight of the riser is decreased by the action of the thermit sleeve below that of the ordinary riser the excess of heat

content over the solidus temperature will also drop

$$\Delta Q = Q_{or} - Q_{red}, \qquad (169)$$

where Q_{red}=heat content of the reduced riser, kcal.

The amount of heat introduced into the reduced riser by the combustion of the thermit mix is

$$\Delta Q = Aq \, G_{mix}, \qquad (170)$$

where A=efficiency of the thermit mix;

q=specific heat of the exothermic process, kcal/kg;

G_{mix}=weight of the thermit mix to heat the metal in the riser, kg.

Equating the right sides of Eqs (169) and (170) we shall obtain

$$Q_{or} - Q_{red} = Q_{or} - Q_{or} \frac{Q_{red}}{Q_{or}} = AqG_{mix}. \qquad (171)$$

Denoting the ratio $\frac{Q_{red}}{Q_{or}}$ through β we shall get

$$Q_{or} = (1 - \beta) = Aq \, G_{mix}. \qquad (172)$$

Substituting the right side of Eq. (169) for Q_{or}, let us solve Eq. (172) with respect to the weight of the sleeve mix

$$G_{mix} = \frac{G_{or} (1-\beta) [c' (\theta_p - \theta_s) + \varrho]}{Aq}. \qquad (173)$$

The value β is given in this equation, and the value A is found experimentally (0.40-0.45). After substituting other constant values we obtain a simple equation for determining the weight of the mixture (thermit and cement in a dry mass)

$$G_{mix} = 0.25 \, G_{or} (1 - \beta), \qquad (174)$$

where G_{or}=weight of the riser as calculated by one of the above methods.

In our experiments the coefficient β was about 0.50-0.55.

The size of an ordinary riser obtained from Eqs (148) through (166) should be decreased by not more than 50-80 per cent for low-alloy carbon steel and by 40-50 per cent for high-alloy steel.

As the mould is being filled the metal enters the riser. Some time after, the thermit mix will ignite and rise to the surface as liquid porous slag while the metal produced by the reduction of Fe_2O_3 will mix with the metal in the riser. The slag forms a layer about 100 mm thick and prevents the metal in the riser from losing heat by radiation.

The metal obtained from ordinary thermit has the following composition: 0.08-0.12 per cent C, 0.5-1.0 per cent Si; 0.0-0.2 per cent Mn, up to 0.05 per cent P and up to 0.05 per cent S.

The properties of thermit mixtures are compared in Table 23.

Table 23

Composition of Thermit Mixtures

Combustible	Thermit composition, %		Thermal effect from combustion of 1 kg of thermit, kcal
	Combustible	Fe_2O_3	
Aluminium	25	75	930
Magnesium	31	69	1,050
Calcium	21	79	930
Silicon	31	69	578

Table 23 clearly shows the advantages and shortcomings of various combustibles used to compose thermit mixtures. Magnesium presents an explosion hazard since it begins to boil and form vapours at liquid steel temperature.

Thermit with silicon often fails to ignite even at 1400° C. Thermit with aluminium burns at 1300° C. Thermits containing calcium, boron and titanium are very expensive. Thermit with chromium has a low pyrometric effect and ignites too slowly. But all these thermits help to maintain steel in a liquid state in the riser and make it serve its purpose much better. Porosity diminishes under the shrinkage cavity. It stands to reason that all the other methods designed to produce an artificial effect on the metal in the riser are very much inferior to heating.

The risers can be heated with an electric arc only in case of large castings or ingots in piece production in view of the cumbersome equipment and greater labour input. The manufacture and mounting of sleeves in no way differ from the usual practice of handling cores which makes this method increasingly popular in many countries of the world.

21. PRESSURE IN THE RISERS

It has been mentioned elsewhere that a temperature difference is needed to move the solidifying metal from the riser to the casting. This is attained by means of pressure difference which is created by the force of gravity or, in terms of hydraulics, by the pressure head.

Let us consider the pressure balance in a casting (Fig. 103) with riser *5* and gate *2*. The metal is fed to the riser through a pouring cup, a sprue and a horn gate *3*. The riser is provided with a pencil pop-off vent *4*. Let us assume that soon after pouring is completed the metal solidifies and forms a primary impermeable

metal skin inside which liquid metal will persist for some time. In conformity with the law of communicating vessels, a pressure difference of the liquid will be set up in the casting and the riser. Since the head above gate *2* in the casting will be higher than in the riser the metal will flow from the higher level in the casting into the riser and a shrinkage cavity will develop in the casting.

If a sand core with a through hole along its centreline is inserted into the riser the interiors of the casting and the riser will communicate with the atmosphere.

Let us further assume that a certain vacuum will be built up due to the volumetric contraction of the freezing metal in the casting and the riser. In this case the atmospheric air will press against the metal in the riser through the hole in the core and will
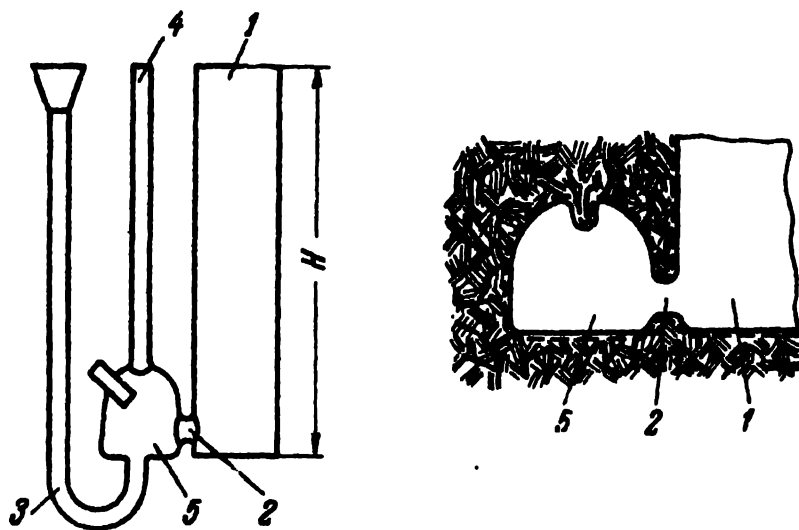
Fig. 103 Casting fed by a riser with an atmospheric core

force the liquid from the riser into casting *1* through gate *2* as in a mercury pressure gauge. The shrinkage cavity will develop in the riser.

These risers were called atmospheric. The idea underlying their operation was developed still further. A gas producing chalk charge was placed into the riser to create an artificial gas pressure on the metal in the riser and force the liquid from it into the casting at 2-3 atm.

Feeding metal with atmospheric and pressure risers was tested by P. I. Yamshanov (U. S. S. R.) and almost at the same time by American specialists. A careful verification of this method undertaken at Soviet and foreign plants showed that neither atmospheric nor pressure risers could provide reliable feeding. In the fifties, fewer and fewer foundries continued to use this method, which is almost completely forgotten now.

Experiments on die-forming of liquid steel in hydraulic presses have shown that an ideally solid structure of metal can be obtained in steel castings not at 2-3 atm but at considerably higher pressures running into 500-600 atm (gauge).

22. MOULDING AND MOULD ASSEMBLY

Mould assembly is the most labour-consuming and important operation in foundry practice, requiring good packing of moulding sands, accurate jointing of the mould elements and efficient feeding both during pouring and the contraction of the metal. All these requirements depend in a large measure on the temperature of liquid steel and the pouring rate through the stopper ladle taphole. All other conditions being equal, the labour input in moulding and assembly operations for steel castings is 10-15 per cent higher than that required to make iron castings.

The intermittent operation of steel furnaces melting a great variety of steel grades with various properties makes the moulds stand for a long time idle and reduces the effective production areas as compared with other foundry processes.

Hence the drive to replace manual labour by mechanised operations and increase the electric power capacities in terms of kilowatt-hours per one rate-hour of the moulder and core maker. The working place of the moulder is organised in a steel foundry in the same way as in other foundries. Below we describe, by way of example, the moulding of steel castings by some of the progres-

Fig. 104. Mould for casting a ram (Vasilyev, Chepel):

1-3—sprues

sive methods. The case in hand will be illustrated by casting a steel ram whose plan and longitudinal section are shown in Fig. 104. The casting made of grade 50Л steel is tested at 450 atm during the hydraulic test. The metal is fed through two sprues 100 mm in diameter. One sprue feeds the metal from the bottom at the start of pouring after which the mould is refilled through the other sprue and six shower gates 50 mm in diameter. The hazard of heating the lower flange involving undesirable negative temperature gradient is eliminated by means of a grid of welded chills made from iron rod 25-30 mm in diameter, which are supported in the mould by chaplets. Thanks to the presence of the chills one riser arranged at the top of the mould will be enough. The riser diameter at the base is 900 mm and at the top 990 mm and it is 1,100 mm high. The net weight of the casting is 16.5 tons, the weight of the chips is 4.5 tons, of the gates 1.2 tons, of the risers 6 tons and of all liquid metal 28.2 tons.

Shrinkage of the patterns is taken at 1.5 per cent and that of the core box 1 per cent. The mould consists of four moulding boxes $2.4 \times 2.4 \times 0.8$ m in size with the drag provided with a recess for a stay bolt 60 mm in diameter and 2,300 mm long which secures the core in the mould. The gases are discharged from the core through a layer of used sand 100-150 mm thick and a pipe 80-100 mm in diameter. The core is reinforced with a grid (Fig. 105). The core is made in a box consisting of two halves and rammed in a horizontal position.

A layer of facing material, 50-60 mm thick and composed of 50 parts of refractory clay, 30 parts of silica sand and 20 parts of used sand is rammed into the mould. The moisture content of the mixture is 7-8 per cent, green compressive strength 0.7-0.8 kg/cm², dry tensile strength 1.5-2.5 kg/cm² and permeability 70-80. Iron hooks and rods are placed around the pattern at a distance of 250-300 mm from it. Then, 100 mm dia pipes for sprues and 50 mm dia pipes for shower gates are placed into the mould. The pipes are secured by the hooks. The runner is connected to two sprues. After the pattern is removed the mould is sprigged and surfaced with a coating compound for the first time before and for the second time after drying. The wash is composed of 57 per cent marshalite and 43 per cent sulphite liquor; the specific gravity of the wash is 1.4-1.45.

The mould and the core are dried for 12 hours at 400° C to a depth of 100-120 mm in a stationary drying oven. The mould is assembled in a permanent moulding pit. Lifting hooks in the core are closed with sand and the core is dried with a gas torch. The stay bolt is inserted before placing the core in its seat. All gaps are carefully closed up with moulding sand, after which grid chills blown with a sand blaster are placed into the mould. Some liquid clay is then applied to the core seat, the core is lowered and secured

with the nut on the stay bolt. The mould is blown with air to remove dust and sand. The upper part of the mould is fitted onto the flask pins after coating the joint with liquid clay. The gaps are closed up. The riser and the pouring cups are covered with iron sheets.

Castings for large blades of hydraulic turbines have enormous surfaces, 6 m long and 3 m wide. Each blade cast in a two-part mould placed on the long edge swelled under the weight of 250 tons of metal and required 1,000 hours to remove the burning-on
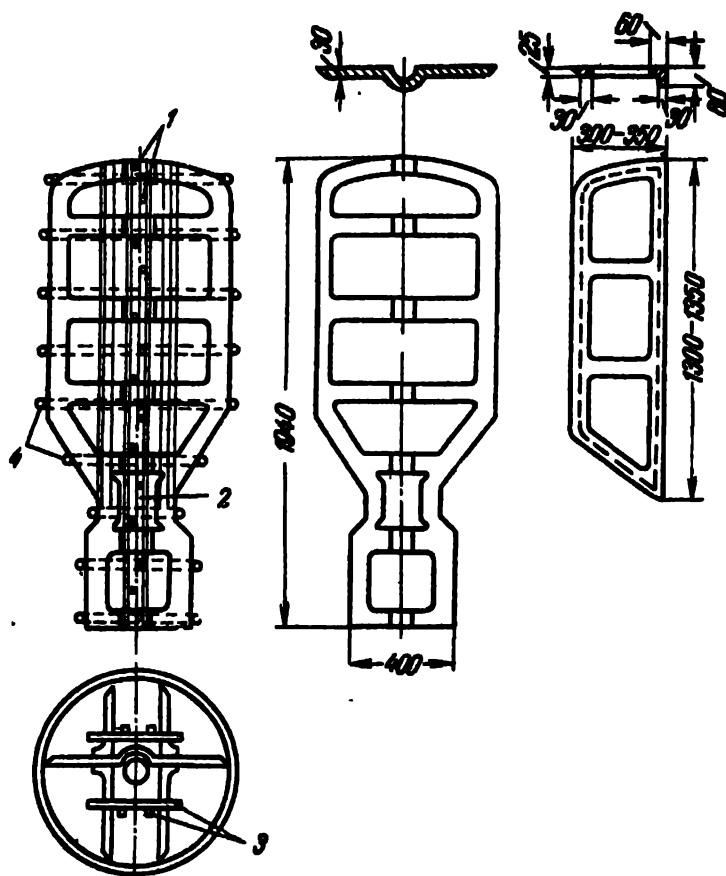
Fig. 105. Grid of the core for casting a ram (Vasi-lyev, Chepel):

1—20 mm dia holes for discharge of gases; 2—100 mm dia pipe; 3—35 mm dia iron rods; 4—20-25 mm dia rods

and about 2,000 hours for size finishing during dressing. A new process was developed and applied in a mechanised permanent moulding pit (Fig. 106) 7.2 m long, 4.95 m wide and 2.25 m deep. Two supports with stationary 1 and movable 5 frames were provided in the bottom of the pit. The movable frame travelled on trolley 4 with runners and plates braced together by studs. The movable part of the frame travelled on two guide rollers 8 which moved in the grooves of the stationary frame.
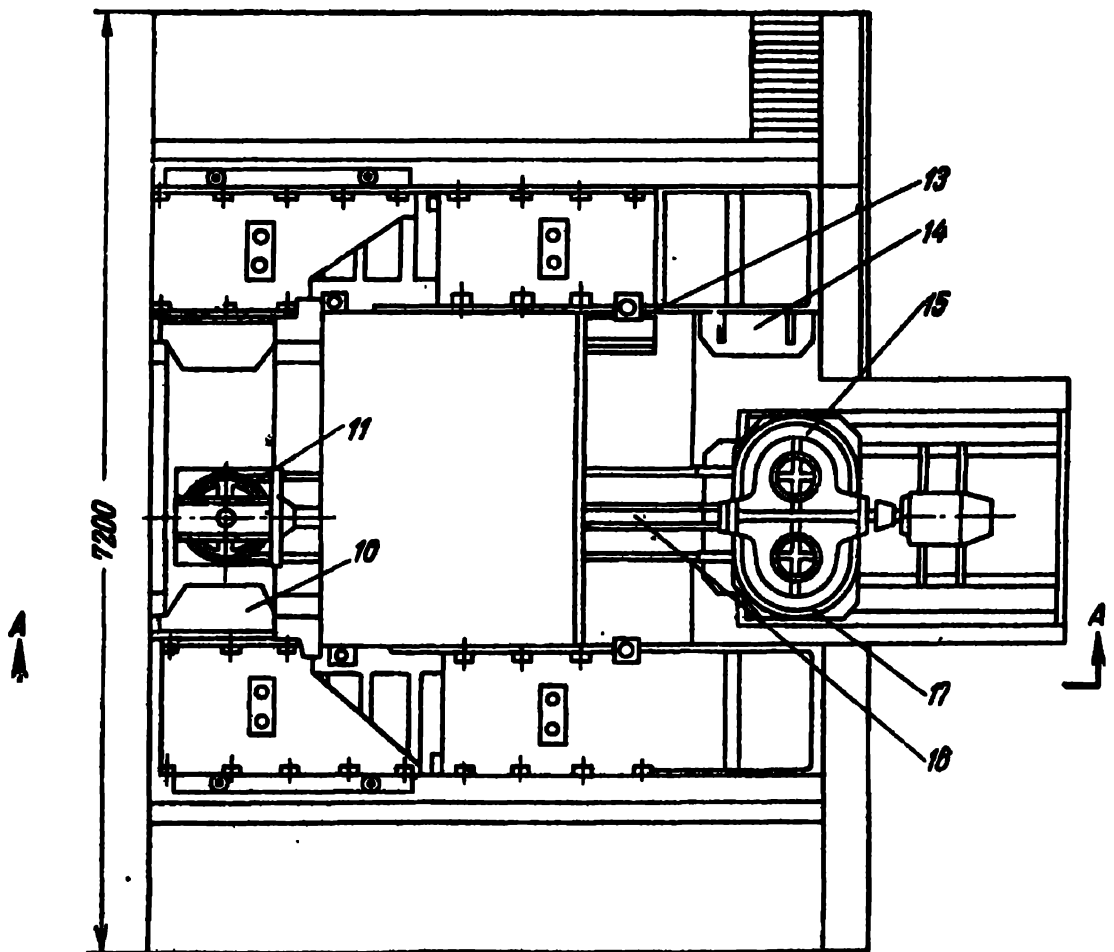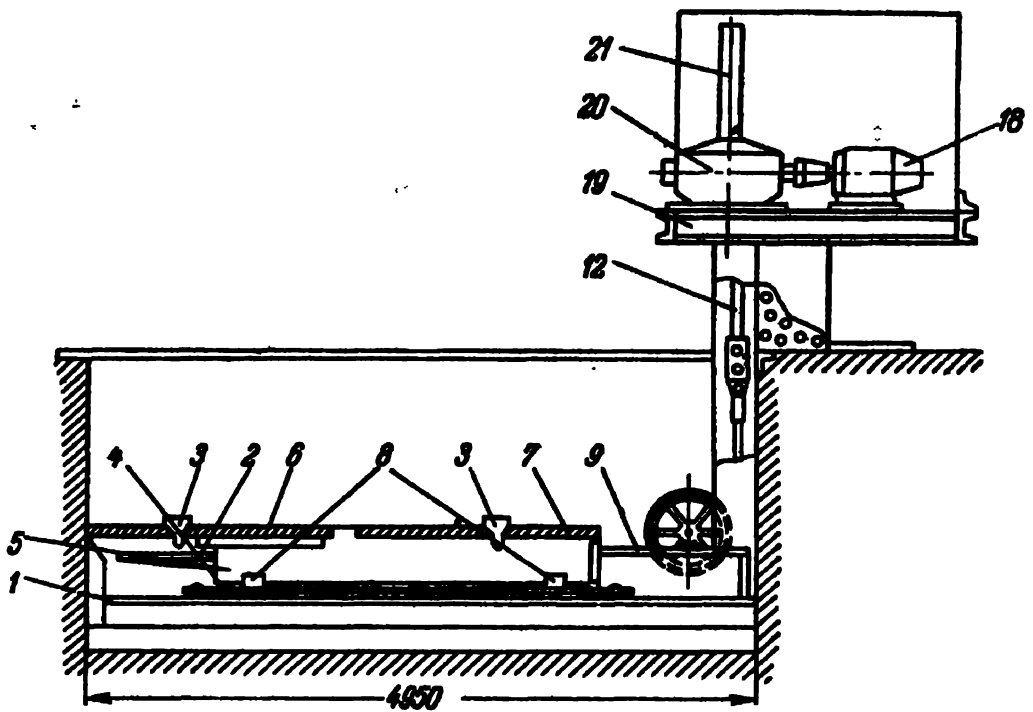
Fig. 106. Permanent moulding pit and mould for the blade of a
hydraulic turbine

Shields 9 in the front part of the support and shields 2 at the rear protected the trolley track from moulding sand. Plates 6 and 7 with rests 3 were secured on the movable and stationary frames to position the moulding boxes, to open and close its halves. The lower frames were connected by welded girders 10 and 14 and the movable frames by girder 13. The two halves of the mould were shifted relative to each other by two wire ropes which connected girder 13 with reducing gear 20. Wire rope 12 was wrapped around sheaves 11 and 17 and fastened to one screw and the other rope around sheave 15 to the second screw of the reducing gear. Spacer 16 was made thicker and girders 10 and 14 were strengthened. Reducing gear 20 and electric motor 18 were mounted on a vertical frame secured on girder 14, and connected by a friction coupling located on platform 19.

Upon rotation, the worm of the reducing gear lifted one and lowered the other screw which by means of wire ropes 12, moved girder 13 rigidly connected with the movable frames of the supports and brought the mould halves apart or together. The reducing gear screws could separate the halves by 800 mm which was quite enough for removing the pattern. The movement of the reducing gear screws was limited by arms 21 and end stops.

Each half of the mould had a vertical joint and consisted of a drag and three copes. The middle moulding box could be removed in case of blade of a different size was cast.

Moulding was begun by securing the metal pattern on props 3 and 7 mounted on the stationary frame (Fig. 107). At the top, the pattern was fastened to moulding box 1 with crosspiece 12, collars 11 and bolts 10. Props 3 and 7 comprising body 4, bar 8, rod 13, pipe 14, a nut and a cotter pin could also be used to separate the pattern from the mould.

Secured at the bottom of the permanent moulding pit were rests 5 with two box beams 6 which entered grooves in the moulding boxes and guided them as they were brought apart or together. The moulding boxes were braced by clamp bolts 2. The refractory walls of the pit were lined with metal plates and the corners with L-iron 9.

The shields were placed at the sides of the pattern after it was secured on its props and at the top. The mould was separated into two parts on both sides of the pattern. Packing was done simultaneously in both parts with pneumatic rammers in the spaces between the side vertical walls of the moulding boxes and the pattern. The packing density was minimum 80-90 by the hardness tester. The pattern was lined with a layer of facing material 20-30 mm thick. The backing sand was packed in horizontal layers not over 100-120 mm thick. 400-mm hooks were positioned at intervals of 200 mm. Saw dust was placed after every 2-3 layers to faci-
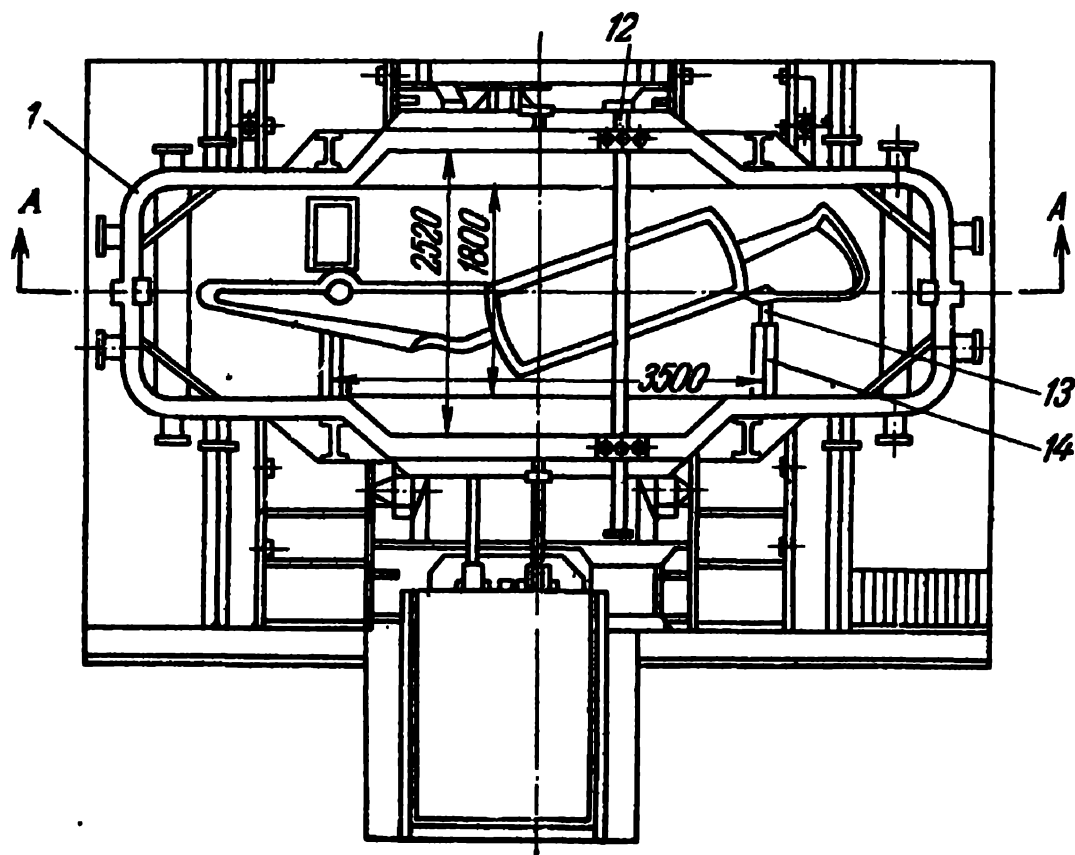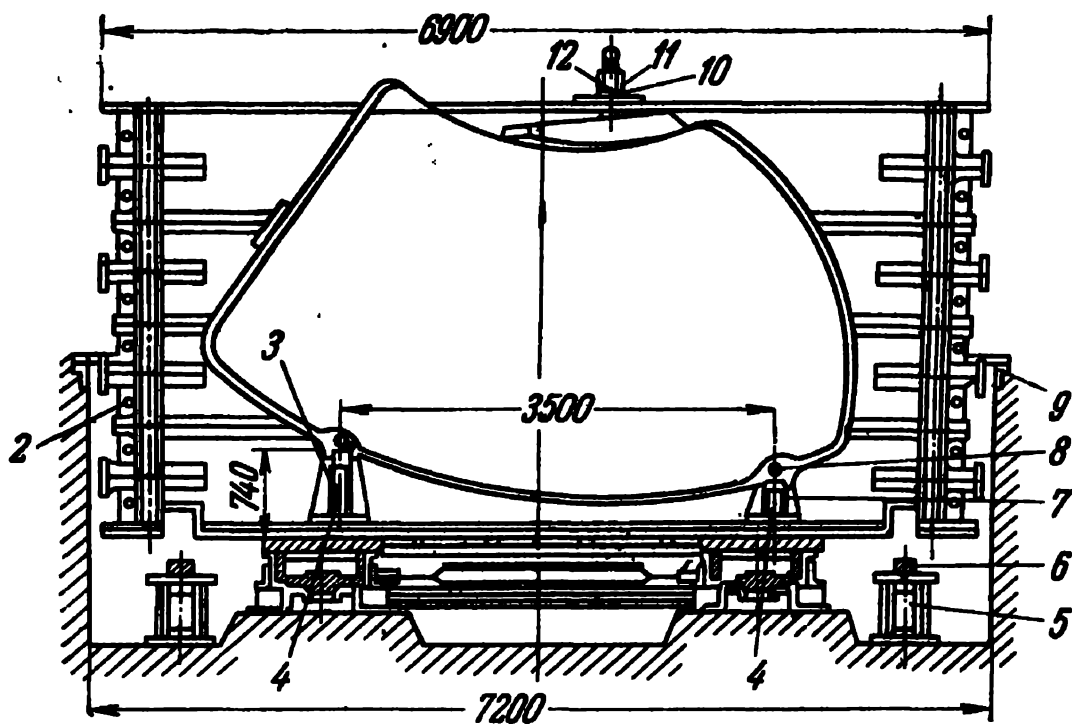
Fig. 107. Mould with a secured metal pattern of a hydraulic turbine blade (Kryanin, Dubrovsky)

litate the discharge of gases to the vent inserted in each half of the mould.

After the lining reached the flanges the upper girder securing the pattern was removed, the flange pattern installed and the mould was rammed to its top, after which the horizontal plane was finished and the runner bush and the pattern of the riser were erected. Two pop-off vents were provided at the upper parts of the blade to vent off the air.

The mould was separated and the pattern removed in the following sequence. At first, the riser pattern was removed with a crane. The runner bush of the riser and the upper moulding box were then removed, turned over and finished. The flange pattern
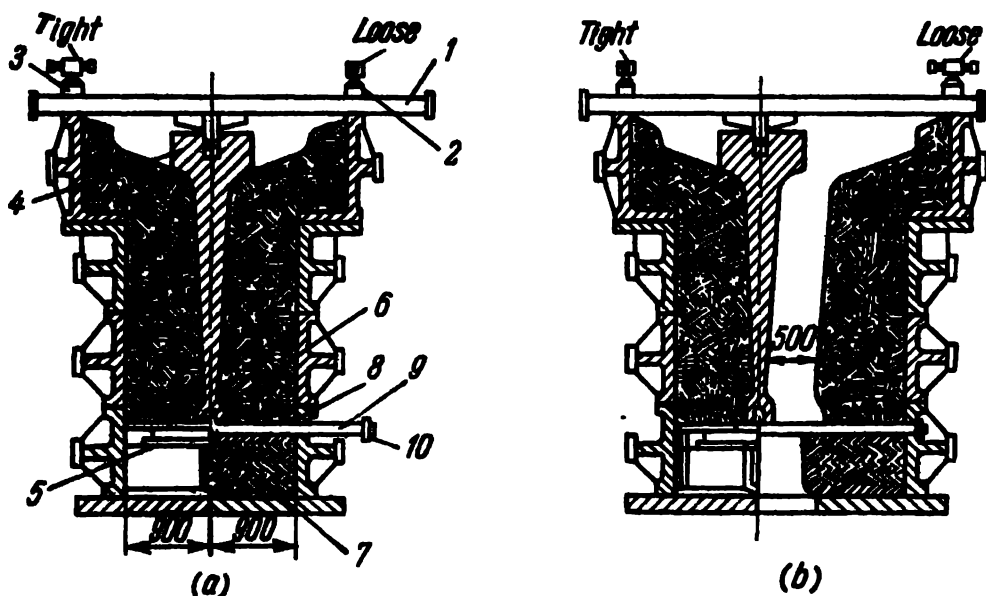
Fig. 108 Ready mould for casting hydraulic turbine blades (Kryanin, Dubrovsky):

a—jointed mould; b—disjointed mould

was taken away, the upper girder mounted and the blade pattern attached to it.

Fig. 108a shows schematically the mould for the blade after the flange pattern was removed and the main pattern was ready to be taken out of the mould. Girder 1 is fastened by screw 3 to the stationary half of the mould and screw 2 is slack. As the electric motor is switched on, the right half of the mould shifts to the right, while pattern 4 and the left half remain stationary. When the right side of the mould is pulled 500 mm away from the left half (Fig. 108b), nut 10 is moved on the thread of rod 9 enclosed in pipe 8 all the way to the moulding box wall. The electric motor is switched on again and the right half of the mould entrains the pattern. For this, screw 2 is tightened and screw 3 is loosened. Then, screw 3 is tightened again. The pattern is attached to the crane hook by a

wire rope. Girder *1* is detached from the mould and lifted up together with the pattern. After studs *6* secured with cotter pin *7* are removed, girder *5* enters into the supports and cores *9* are pulled out of the mould. The holes in the mould are closed up with a mixture. After finishing the mould, both halves are brought together to a clearance of 440 mm covered by the shields and are dried by portable driers.

The dried mould is assembled and installed in the permanent moulding pit for pouring. The horizontal plane of the metal plate provided in the bottom of the pit is machined with cutters. The bottom surface of the plate is likewise machined to obtain a full contact between the mould and the plate and prevent leakage of metal from the mould.

The mould is secured by wedges in the pit and clamped with screw jacks which prevent its swelling and deflection. This method has decreased the deformation of the mould 6 to 8 times and increased labour productivity 4-5 times.

23. EXTERNAL AND INTERNAL CHILLS. APPLICATION AND CALCULATION

Apart from decreasing the yield of good ingot and prolonging the time needed to remove chips, shrinkage allowances involve considerable difficulties in moulding since they may become opposite in direction to the drift angle required to assist the withdrawal of the pattern from the mould. This may be observed in moulding shaped castings illustrated by a part for a steam turbine in Fig. 109. Wedge-shaped shrinkage allowances are out of the question here. The freezing of metal can be artificially promoted by the use of chills to ensure directional solidification and prevent the formation of centreline porosity in the heat centres far from the risers. External chills are placed in the mould walls at the mould-metal interfaces. Internal chills are placed in the mould cavity.

External chills are used to cool such projections as bosses and lugs (Fig. 110, left) or else in the reentrant L-shaped angle with a thickened section (Fig. 110, right).

The standards for external chills were suggested by many Soviet scientists, including Y. Nekhendzi, A. Arkhangelsky, V. Desnitsky, P. Vasilevsky, V. Bugachev, A. Skvortsov, who developed different calculation methods illustrated in the summary Table 24.

V. Bugachev, who investigated the effect produced by the duration τ of the steel flow in the mould past a chill and the metal superheat in pouring, deduced approximate equations of heat balance of the chills depending on the geometrical factor of the centre being cooled (the height-to-radius ratio of the centre $M = \frac{H}{R}$

or $\frac{\varrho}{R}$). These equations served the basis for compiling an alignment chart (Fig. 110, bottom) which is used to establish the thickness of a triangular or plate chill. For heavy sections with $\frac{H}{R}$ from 1 to 2, two-way cooling should be provided and for less than 0.2 no
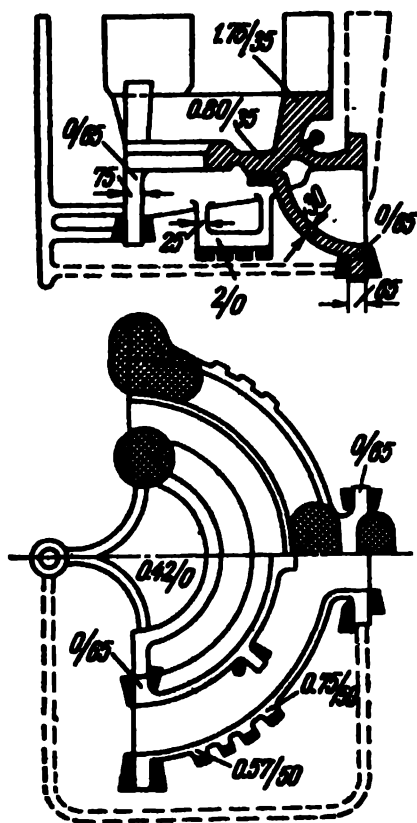
Fig. 109. Local chills placed in the mould for casting a part of a steam turbine (Bugachev)

chills are required at all. For reentrant angles, a fillet of radius ϱ can be recommended with $M_{1r}=\frac{\varrho}{R}$ from 1 to 2.

Example 1. Find the thickness of a chill for a flat casting with a wall thickness of $2s=40$ mm, a length of $l=80$ mm and with a boss $h=20$ mm thick and $d=30$ mm high. The height-to-radius ratio of the centre is

$$M = \frac{H}{R} = \frac{hd\,(2s+l)}{(4h+d)\,ls} = \frac{20\times30\,(40+80)}{(4\times20+30)\,80\times20} = 0.4.$$

The duration of the steel flow through the centre is $\tau=60$ sec. Pouring is done at a low superheat, about 5 per cent. To find the thickness of a plate chill, let us draw on the alignment chart (Fig. 110) a straight line from point 0.4 on the left vertical to intersection with the left vertical on the right side of the alignment chart at point 60. The result will be the point of intersection of the inclined straight line with the left central vertical at the point which corresponds to the maximum thickness of the

Fig. 110. Alignment chart for determining the size of external chills by the given steel superheat and the duration of steel flow past the chills (Bugachev)

Table 24
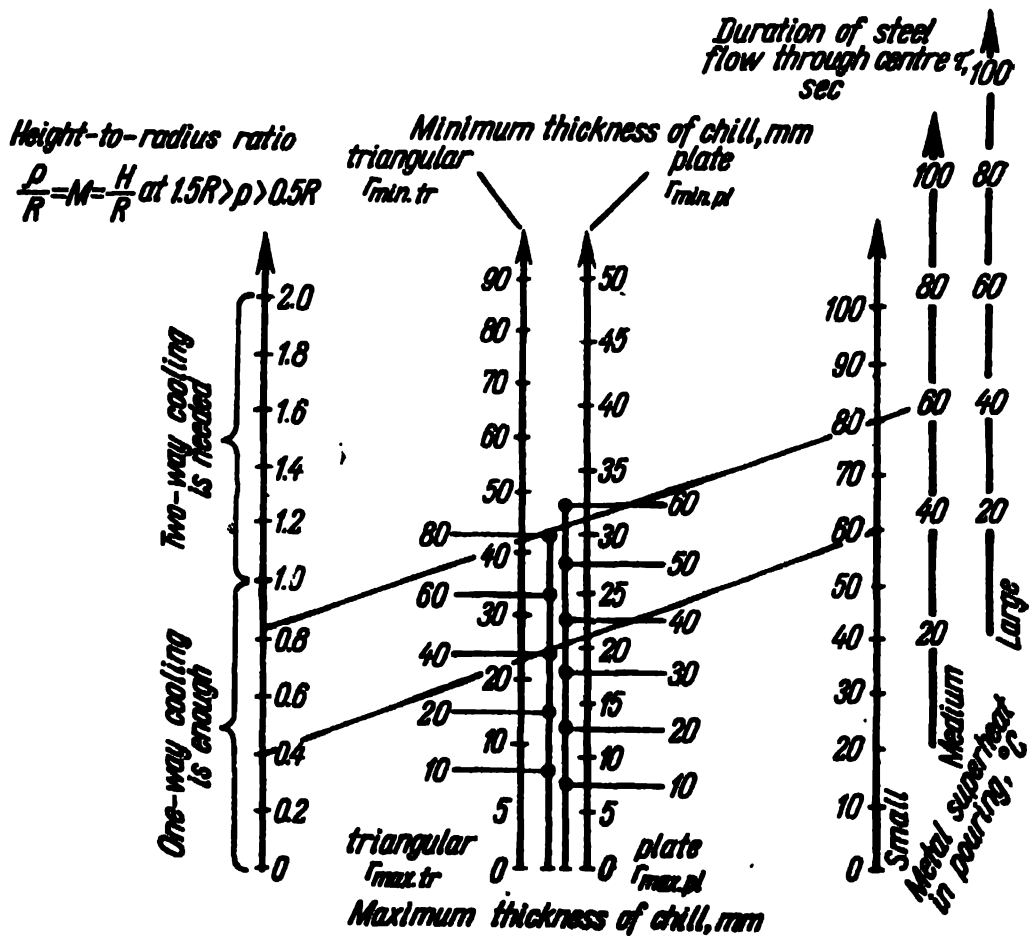
Standards for External Chills

(see Fig. 110)

Calculation method		Group of castings by the thickness of the main wall, cm			Authors
		2s < 1	2 > 2s > 1	5 > 2s > 2	
By thickness of main wall		—	$2.0 > \dfrac{r}{2s} > 1.5$	$1.5 > \dfrac{r}{2s} > 1.0$	Nekhendzi
By thickness of main wall		$r = 2s$ to $3s$	$r = 1.6s$ to $2.4s$	—	Arkhangelsky
By nature of heat centre — By flange thickness	1.5-3.0	$r = 0.9$ to 1.4			Desnitsky
	3.1-5.0		$r = 2.0$ to 2.8		
	5.1-7.6			$r = 3.6$ to 4.8	
By projection thickness	At $s > h$			$r = 0.6h$ to $1.0h$	
By diameter of circle inscribed in centre being cooled	1.2-3.0	One-way cooling $r = 0.6D$ Two-way cooling $r = 0.3D$ One-way cooling $r = 0.1D$ Two-way cooling $r = 0.4D$			Vasilevsky

· Notation: 2s—mean thickness of the casting main wall, cm; h—thickness of projection on the casting, cm; r—chill thickness, cm; ρ—radius of the reentrant angle fillet, cm; R, D—radius and diameter of the circle inscribed in the centre being cooled, cm; l—projection length, cm; H—referred thickness of the projection, cm.

chill—35 mm and with the right central vertical line at the point which corresponds to the minimum thickness of the chill—21 mm.

Example 2. Find the thickness of a triangular chill for an L-shaped casting (Fig. 110, right) with a thickness of $2s=60$ mm, length of $l=150$ mm and fillet radius 20 mm. The height-to-radius ratio of the centre is

$$M = \frac{\varrho}{R} = \frac{\varrho\,(2s+l)}{ls} = \frac{20\,(60+50)}{150\times30} = 0.90.$$

Pouring is done at a medium superheat equal to 10 per cent. The duration of the steel flow through the centre is 60 mm/sec. Let us connect point 0.90 on the left central vertical with point 60 on the middle right vertical. We shall obtain the maximum point 80 mm and the minimum point 43 mm at the intersection with the middle left verticals.

The chills for the elements of a steam turbine (Fig. 109) can be calculated by the alignment chart as shown in Table 25.

The duration of solidification of individual centres can be determined from P. Berg's approximate equation

$$\tau_{sol} = (8 \pm 2)^2\,R. \tag{175}$$

The maximum and minimum thickness of a plate chill can be found from the equations

$$r_{max} = \frac{\alpha\tau_{fl}\theta_p + (R+H)\,q}{\gamma_{ch}c_{ch}\left[\theta_s - \dfrac{q(R+H)}{\alpha t_{sol}}\right]}\,; \tag{176}$$

$$r_{min} = \frac{\alpha\tau_{fl}\theta_p + (R+H)\,q}{\theta_{f.ch}\,\gamma_{ch}\,c_{ch}}\,, \tag{177}$$

where $\alpha=$ heat conductivity coefficient of a chill coated with a thin layer of marshalite compound, found experimentally and equalling approximately 0.03 cal/cm^2 sec °C and for a thicker layer of coating 0.003 cal/cm^2 sec °C;

θ_p, θ_s and $\theta_{f.ch}=$ pouring, solidus and fusion temperatures of the chill respectively;

$q = \gamma_{ch}\,[\,W + c_{ch}\,(\theta_p - \theta_s)]=$ amount of heat absorbed by the chill, cal;

$\gamma_{ch}=$ specific weight of the chill;
$W =$ latent heat of fusion;
$c_{ch}=$ heat capacity of the chill.

The chill must heat up but not fuse, and therefore the value of q should have its limit.

Table 25

Calculation Data for Steel Casting Chills (Bugachev)

Designation in Fig. 109	Heat centres						Ratios of		Duration of metal flow t, sec	Design thickness of chills, mm		
	Centre	s, mm	$2R$, mm	h, mm	d, mm	H, mm	thickened section $\frac{H}{R}$	angle $\frac{R}{Q}$		near gates	far from gates minimum	far from gates maximum
0.42/0	Lug-rib	30	30	10	70	6.3	0.42	—	0	—	7	17
0.57/50	Rectangular boss . . .	30	30	22	65	8.5	0.57	—	50	—	19	24
0.75/50	Round boss	30	30	32	70	11.0	0.75	—	50	—	20	36
0.80/35	Local thickened section	30	30	30	80	12.0	0.80	—	35	—	17	36
2/0	Bracket lug plane . . .	25	25	30	230	25.0	2.00	—	0	—	21	Two-way 36+36
0/65	Small flange	65	41	0	100	0	0	—	65	23	—	—
1.1/0	Same, without riser . .	30	30	40	100	16.7	1.00	—	0	—	25	Two-way 36+36
0;65	Small flange.	75	50	0	180	0	0	—	65	26	—	—
1.75/35	Reentrant angle $Q = \doteq 20$ mm	70	70	—	—	—	—	1.75	65	40	—	—

The thickness of a triangular chill can be found from the following equations

$$r_{tr.max} = \frac{\frac{\pi}{2} a\tau_{fl}\theta_p + Rq\left(2+\frac{R}{q}\right)}{\sqrt{2}\,\gamma_{ch}c_{ch}\left[\theta_s - \frac{2Rq\left(2+\frac{R}{q}\right)}{\pi a\tau_{sol}}\right]} - \frac{\varrho}{2}; \qquad (178)$$

$$r_{tr.max} = \frac{\frac{\pi}{2} a\tau_{fl}\theta_p + Rq\left(2+\frac{R}{q}\right)}{\theta_{f.ch}\sqrt{2}\,\gamma_{ch}c_{ch}} - \frac{\varrho}{2}, \qquad (179)$$

where $H = \frac{hd}{4h+d}$ = equivalent of the chill thickness;

$R = \frac{ls}{2s+l}$ = equivalent of the main thickness of the casting.

When the heat centre is T-shaped (Fig. 111) the two-way cooling becomes insufficient and a third chill is required. Its height h should be from 1.2 to 1.5 of the thickness of the casting main wall.

An X-shaped centre (Fig. 112a) should be avoided as much as possible, since even four triangular chills will not be able to ensure a sufficiently solid metal in the heat centre. It is frequent practice to replace an X-section by two T-sections (Fig. 112b).

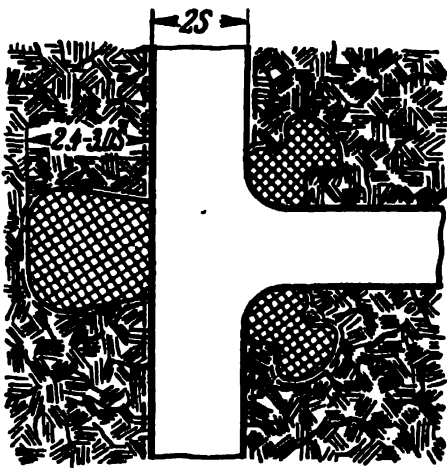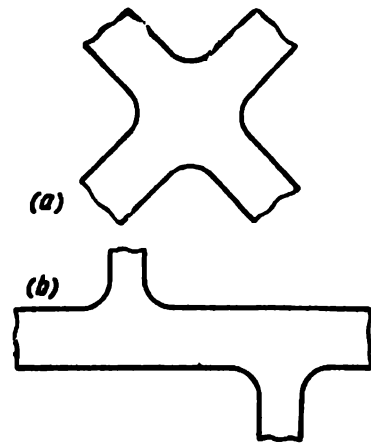*Fig. 111.* Location and shape of external chills for two-way cooling of a T-joint

Fig. 112. The use of two T-sections (b) instead of one X-section (a)

When the moulds are filled at a low superheat, shrinkage ribs which will act as chills for the heavy section can be used instead of triangular chills.

The maximum thickness of a casting to be chilled with the external chills of adequate thickness in one-way cooling (Fig. 111)

can be found from the equation

$$\frac{s+h}{s} = 1.7 \quad \text{to} \quad 2.0, \tag{180}$$

in which case h should not exceed (0.35 to 0.50) $2s$. In two-way cooling the thickness of the chill should be doubled.

Internal chills are arranged inside a thickened section of the casting; they absorb part of the heat of the solidifying metal, like external chills. Internal chills are intended primarily to compensate for the harmful effect of the negative longitudinal temperature gradient, such as occurs, for example, at bottom gating (see Fig. 44) and when the riser is so small that its inscribed circle is smaller than that of the underlying thickened section of the casting. The bottom part of the casting, like L- or T-shapes, can be cooled faster by means of two- or three-way cooling only when the ratio of the inscribed circles, $D : d$, does not exceed 1.5 to 1.6. If this ratio exceeds 2.5, internal chills should be used.

The weight of the internal chills comes to 5-8 per cent of the weight of the section being cooled. If the chills are too light nothing can prevent the formation of a shrinkage cavity. Contrariwise, if the internal chills are too heavy they will neither fuse on the surface nor weld with the parent metal of the casting. This will give rise to porosity with stresses concentrated around them.

If not cleaned to glitter with sand or shot blasters internal chills with moisture from the mould condensed on them, will become rusty and promote the formation of blowholes.

The use of internal chills is therefore limited by certain conditions. The total weight of the chills should be selected with account taken of their heating to the fusion point of the surface. The size and weight of the chills can be found with the help of the heat balance equations.

To fuse a chill completely, the pouring must be done at a high superheat (over 10 per cent) but in this case the feeding of the casting in the space between the chills will be impossible to achieve. The pouring temperature should be such that the time of solidification of the upper part of the casting is slightly above the time required by the bottom part to solidify

$$\tau_{up} > \tau_{bot}. \tag{181}$$

To eliminate centreline porosity between the chills the thickness of the metal layer increasing during solidification δ_{ch} should be slightly above the thickness of the layer which develops on the mould wall, δ_m, i. e.,

$$\delta_{ch} > \delta_m. \tag{182}$$

Comparison of Eqs (181) and (182) shows that the growth of the layer δ_{ch} will be more intensive around a cylindrical chill than around a flat plate and, from a theoretical point of view, a cylin-

der or even a ball will be a more advantageous form for the internal chill than a square or plate.

The complexity of heat calculations and their insufficient accuracy made it impossible to determine correctly the size of internal chills and frequently limited their application, despite their marked superiority over external chills. Their size was often established empirically. An attempt to determine the size of internal chills for the ratio between the thickness of a square or round casting d_c and the thickness of a thin riser d_r of 2.0 to 2.5 was made only recently.
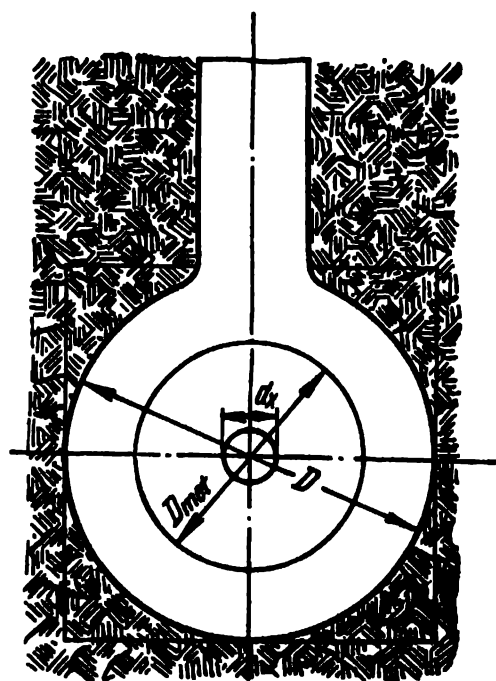
Fig. 113. Calculation of the size of an internal chill d for a cylindrical thickened section D (Skvortsov)

Below are given the results of calculations based on the determination of the time needed by the riser and casting to solidify, according to Eq. (175), and depending on the degree of the steel superheat during pouring (Fig. 113).

A. Veinik's equation is used to find the diameter of the internal part of the casting D_{met} which solidifies due to the heat absorbed by a chill with a diameter d_{ch}, in conformity with the equation

$$d_{ch} = D'_{met} \sqrt{\frac{\gamma_1 [c_1 (\theta_p - \theta_s) + Q_1]}{\gamma_{ch} c_{ch} (\theta_{f.ch} - \theta_{i.ch}) + \gamma_1 [c_1 (\theta_p - \theta_s) + Q_1]}}, \quad (183)$$

where γ_1 = specific weight of liquid steel, g/cm³;
c_1 = heat capacity of liquid steel, cal/g °C;
Q_1 = latent heat of solidification, cal/g;
γ_{ch} = specific weight of the chill, g/cm³;
c_{ch} = heat capacity of the chill, cal/g °C;
θ_p, θ_s = pouring and solidus temperatures, °C;
$\theta_{f.ch}$ and $\theta_{i.ch}$ = temperature of the chill at the end of heating (final) and its initial temperature, °C.

Let us assume that half of the rod is fused. Then, with the rod diameter d_{ch}=0.01 m the diameter of a metal chill will be

$$D'_{met} = 0.01 \sqrt{1 + \frac{7,700 \left[0,168\,(1,450-20)+\frac{1}{2}\,65\right]}{7,100\,[0.19\,(1,550-1,475)+65]}} =$$
$$= 0.022 \text{ m} = 22 \text{ mm}.$$

The effect produced by the pouring temperature θ_p on the diameter of a cylindrical chill (Fig. 113) can be found from Eq. (177): in the absence of superheat

$$d_{ch} = 0.445\,D'_{met}\,; \tag{184}$$

with a 50° superheat

$$d_{ch} = 0.470\,D'_{met}\,; \tag{185}$$

with a 100° superheat

$$d_{ch} = 0.500\,D'_{met}\,. \tag{186}$$

The dimensions of internal chills are illustrated for different cases of this casting for the ratio of $d_c : d_r$=2.0 in Table 26.
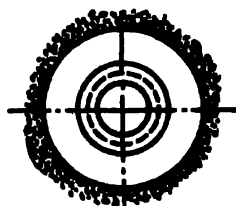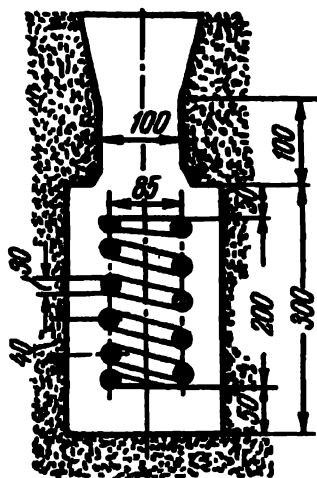
Fig. 114. Calculation of a coiled chill for a cylinder with a small riser (Skvortsov)

An internal coiled chill intended for a cylindrical casting with a small riser (Fig. 114) cools the casting fairly well and fuses readily enough.

The dimensions of chills for castings shown in Figs 113 and 114 with the ratio $d_c : d_r$=2.5 to 3 are represented in Table 27.

Table 26

Dimensions of Internal Chills for Cylindrical and Square Thickened Sections
(Skvortsov)

Parameter	Thickness of the upper part of the casting, mm								
	20	30	40	50	60	80	100	150	200
Diameter of the lower part of the casting, mm	40	60	80	100	120	160	200	300	400
Solidification time of the upper cylindrical part of the casting, min:									
without superheat	0.0140	0.030	0.056	0.090	0.130	0.287	0.355	0.800	1.41
with a 50° superheat	0.167	0.037	0.067	0.108	0.151	0.270	0.420	0.940	1.67
Solidification time of the bottom part of the casting, min:									
without superheat	0.0130	0.025	0.048	0.077	0.110	0.177	0.290	0.670	1.20
with a 50° superheat	0.0160	0.032	0.060	0.095	0.130	0.220	0.370	0.850	1.50
Chill diameter, d_{ch}, mm:									
without superheat	3.0	8.5	9.5	11.0	17.0	24	26	37	50
with a 50° superheat	4.6	9.5	12.0	14.0	19.0	25	30	48	60
Diameter of the coil wire, d_{ch}, mm	2.0	3.5	4.5	6.0	7.5	10	15	20	—
Coil diameter, D_{ch}, mm	6.0	12.5	16	18	25	34	40	62	—

Table 27

Dimensions of Internal Chills for Castings Shown in Figs 113 and 114
(Skvortsov)

Ratio $d_c{:}d_r$	Parameter	Thickness of the upper part of the casting, mm								
		20	30	40	50	60	80	100	150	200
2.5	Diameter of the lower part of the casting, mm	50	75	100	125	150	200	250	375	500
	Chill diameter, mm:									
	without superheat	11.0	16	22	27	31	42	52	78	109
	with a 50° superheat	12.5	19	26	32	38	50	61	92	127
	Diameter of the coil wire, mm	4	6.5	8	10	—	—	—	—	—
	Mean coil diameter, D_{ch}, mm:									
	without superheat	15	22	30	38	—	—	—	—	—
	with a 50° superheat	4.5	7.5	9	12	14	16	20	—	—
	Same	16	25	34	42	50	65	80	—	—
3.0	Diameter of the lower part of the casting, mm	60	90	120	150	180	240	300	450	—
	Chill diameter, d_{ch}, mm:									
	without superheat	16	25.5	32.5	41.0	49	68	84	120	—
	with a 50° superheat	18.5	31.0	38.0	47.0	58	78	98	140	—
	Diameter of the coil wire without superheat, d_{ch}, mm	6.5	9	11	14	—	—	—	—	—
	Coil diameter, D_{ch}, mm	22	33	45	58	—	—	—	—	—
	Diameter of the coil wire without superheat, d_{ch}, mm	7.0	10	13	16	18	—	—	—	—
	Coil diameter, D_{ch}, mm	24	40	50	60	75	—	—	—	—

The rule of the inscribed circles allows us to employ to a certain approximation the calculation data from Tables 26 and 27 for various cases of shaped castings, i. e., in the presence of local thickened sections in the form of lugs, bosses and stapped shoulders (Fig. 109). The diameter of the inscribed circle of the thickened section is taken as d_c and of the adjoining main wall of the casting as d_r.

Table 28

Initial Data for Calculating a Pack of Internal Chills by the Given Thickness of a Rod and the Degree of Steel Superheat

(Skvortsov)

Parameter	Diameter of the chill rod, mm									
	3	5	7	10	15	20	25	30	40	50
Solidification time of the part of the casting being cooled during which the rod is heated and partially fused, min	0.50	0.75	1.10	1.80	3.0	5.0	6.2	7.5	11.0	15.0
Diameter of the metal layer solidified around the rod D'_{met}, mm:										
without superheat	6.8	11.4	16.0	23.0	34.0	46.0	57.0	68.0	91.0	115
with a 50° superheat	6.5	10.9	15.2	22.0	32.5	44.0	54.0	65.0	87.0	108
with a 100° superheat	6.2	10.4	14.5	21.0	31.0	42	52	62	83	104

The formulas for calculating D'_{met} by the given diameter of an internal chill make it possible to calculate a pack of chills for heavy castings with relatively small risers, for example, for castings for die beds or press plates. The calculations provide for a partial or even complete fusion of the chills.

The rods should be so arranged in the mould cavity as to obtain a layer of metal that would solidify around the rods and close up the entire space of the casting between the adjacent rods at the expense of the heat taken off by the chill pack, irrespective of the size of the thin and thick parts of the casting.

Table 28 illustrates the calculation of a pack of chills for a heavy casting with a small riser (Fig. 115).

The more rods there are in the pack, the more time is required to place them into the mould. For this reason the largest rods should be selected for the pack. However, thin rods promote the homogeneity of metal structure in the heavy section. The rods should be free from rust, oil spots, etc. At some plants the rods for internal chills are chrome-plated or coated with aluminium paint to prevent condensation of vapours. It will be well advised to form medium-size rods into boxes to resemble a space lattice, like the chaplets for the cores, and insert them into the mould after drying.
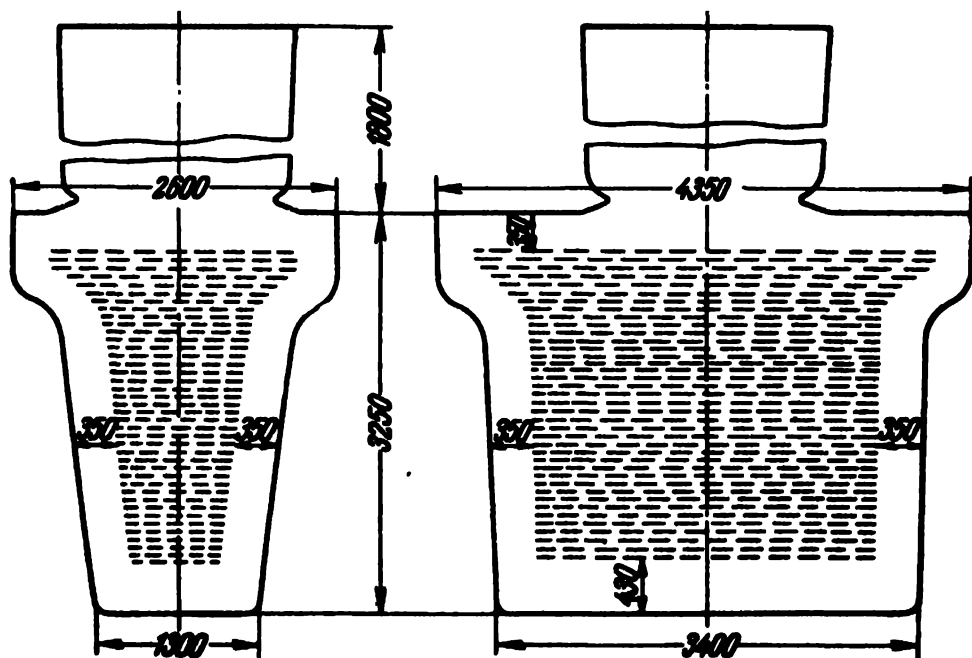
Fig. 115. Installation of a pack of internal chills in the mould for a die bed; rough weight of the casting with the risers is 158 tons, the net weight of the casting is 116 tons

The data in Table 28 show that, when the number of rods in the pack and the diameter of each rod are given, the mould should be filled at the specified superheat. Disregard for this condition prevents the rods from being welded to the parent metal at too low pouring temperature or causes shrinkage porosity to develop in the space between the rods when the steel is excessively superheated.

Example of Calculating a Pack of Internal Chills. A mould for a die bed is shown in Fig. 115. The rough weight of the bed is 158 tons and that of the riser 40 tons. The area shaded in the drawing is filled with packs of chills. Using Table 28, we select iron rounds 30 mm in diameter. The pouring is done at a superheat of 50° above the liquidus temperature, which amounts to 1500° C for steel of grade 20 Л. The layer of the parent metal growing around each rod is 65 mm.

The rods are arranged in the pack as shown in Fig. 116. The layer of metal that has united with the rod is indicated in dash circles. The gaps between them are compensated for by crosswise rods of the same diameter (30 mm) placed at intervals of 180 mm from each other. One cubic metre can accommodate 1.54 tons of lengthwise and 0.51 ton of crosswise rods, a total of 2.06 tons. The space occupied by the grid chills is 5.34 cu m, and the total weight of the chills placed into the mould is $2.06 \times 5.34 = 11$ tons, which is $\frac{11 \times 100}{116} = 9.5$ per cent of the weight of the bed without the riser.
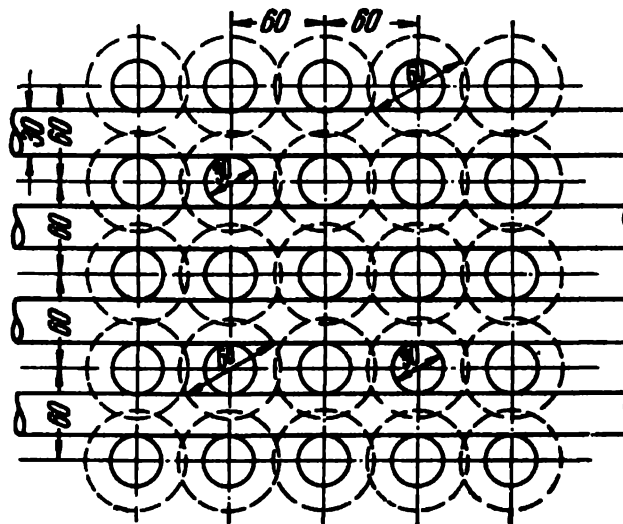
Fig. 116. Arrangement of crosswise and lengthwise rods in a chill pack for a die bed (see Fig. 115)

Due to a much faster cooling from the pack of chills, the weight of the bed riser is about 42 tons or some 36.2 per cent of the casting weight.

According to Briggs, the thickness of the riser wall for a cubic casting should equal 1.1. In our case the diameter of the riser for the bed (regarded as a cube) would be equal to 2,860 mm.

The height of such riser is

$$H_r = \frac{0.7 V}{d^2} + 4 \text{ cm} \quad \frac{0.7 \times 149,000}{286^2} + 4 = 1,280 \text{ mm}$$

and the weight of the riser

$$\frac{\pi \times 2.86^2}{4} \, 1.28 \times 7.8 = 64 \text{ tons,}$$

which is 1.5 times more than that of the riser for the bed with chills. It is quite impossible to detach a riser of such an enormous diameter from the frozen casting.

When the riser is heated with a thermit sleeve its weight can be reduced from 40 to approximately 25-28 tons.

24. NECKED-DOWN RISERS

The separation of risers from steel castings is a difficult and expensive operation which hinders the timely fulfilment of the production schedule. Back in 1911, P. Ivanov designed a riser which could be separated from an ingot of gun steel by using a wafer core provided at the riser neck, thereby greatly simplifying the fettling operation.

During the Second World War, A. Ryzhikov and A. Popov designed, each in his own way, easily separable risers for steel and other castings. In this design (Fig. 117), core 2 made of oil sand
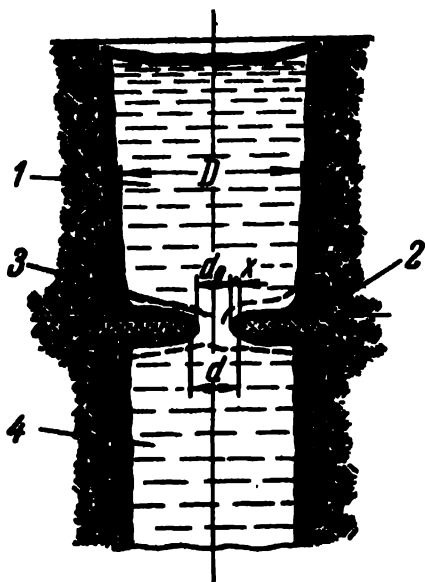
Fig. 117. Design of a wafer core placed into the mould at the neck of the riser

or chamotte clay with centre hole 3 was inserted into the neck of the riser 1, which feeds liquid metal from the riser to casting 4 during contraction. Experience goes to show that in these conditions the riser is no less inferior to an ordinary riser if, of course, it holds enough superheated liquid steel, to the temperature of which the wafer core has been raised and the centre hole is wide enough to feed the casting.

The edges of the cores should be sharpened to facilitate the removal of the risers.

The riser can be easily separated from the casting by striking against a wedge inserted into the gap formed by the core outside the riser. The riser should be separated from the casting soon after solidification. A riser cooled to room temperature can be hardly separated at all.

When a necked-down riser about 1,000 mm in diameter is separated from a die bed weighing 30-40 tons, about 60 cylinders of oxygen, 20 cylinders of acetylene and 30-40 rate-hours can be saved as compared with the separation of an ordinary riser.

Let us consider the operation of the wafer core from the viewpoint of thermotechnics. A certain layer of metal will freeze on

the walls of casting *4*, riser *1* and core *2* at the end of pouring (Fig. 117) and the diameter *d* of the hole will decrease to d_e.

The amount of heat absorbed by the unit surface of the cold core in heating the riser metal to solidus temperature θ_s can be found from the following equation

$$q = \delta_2 \gamma_2 c_2 \theta_s,$$

where $\delta_2 =$ thickness of the core, cm;
$\gamma_2 =$ specific weight of the core, g/cm^3;
$c_2 =$ core heat capacity, cal/g $^\circ$C.

The transfer of this heat from the heated metal will cause the solidification of the layer, *x* thick:

$$\delta_2 \gamma_2 c_2 \theta_s = \gamma_1 Q x, \tag{187}$$

where $\gamma_1 =$ specific weight of the metal, g/cm^3;
$Q =$ heat of solidification, cal/g.

The hole between the cores *d* should be so selected as to ensure easy passage of liquid metal. The general formula for calculating the hole is

$$d = d_0 + \frac{2\delta_2 \gamma_2 c_2 \theta_s}{\gamma_1 Q}. \tag{188}$$

Substituting the numerical values in this formula we obtain

$$d = d_0 + \frac{2\delta_2 \, 2.6 \times 0.15 \times 1,500}{6.9 \times 65}$$

Since $d_0 = 9$ mm, the relationship between the diameter of the hole and the core thickness will be

$$d = (9 \text{ to } 10) + 2.4\,\delta_2 \text{ mm.} \tag{189}$$

At the Urals Engineering Plant the wafer cores are composed as follows: 95.5 per cent silica sand, 2 per cent linseed oil and 2.5 per cent water. The foundrymen at the South Urals Plant use a mixture made of 65-70 per cent sand with 30 per cent marshalite, 10 per cent soluble glass and 1.5 per cent NaOH (10-per cent solution).

The cores are dried at 270-300° C for 4-5 hours. At the Kovrov Excavator Works the cores are formed in a metal die by blows of a heavy sledge hammer or under a press (60-70 per cent crushed chamotte with 30-40 per cent refractory clay). After drying the compressive strength is 38 kg/cm^2 and after baking 61 kg/cm^2.

The observations show that the softened cores tend to sag in moulding large castings under the weight of the metal. Each plant uses therefore its own standards for the core size depending on production conditions. The sizes of wafer cores for heavy castings depending on the diameter of the risers are illustrated in Table 29.

Table 29

Standards for Wafer Cores

(Blokhin and others)

Diameter of riser base, mm	Core size, mm			Feeding hole diameter, mm	Diameter of chaplets, mm	Size of shoulders, mm
	length	width	thickness			
800	1,100	1,100	60	150	15	80-100
1,000	1,300	1,300	70	200	15	80-100
1,150	1,450	1,450	80	300	25	80-100
1,250	1,600	1,600	100	300	25	80-100

One design of a wafer core is shown in Fig. 118. To avoid fusion to the riser or the casting the chaplets should not be positioned in the vicinity of the core. Cracks which appeared after drying should be carefully closed up.
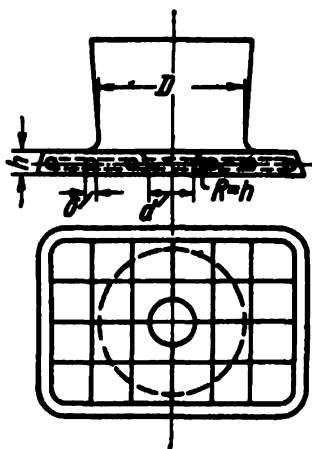
Fig. 118. Wire chaplets for a wafer core in medium and large risers (Blokhin et al)

The chaplet should be placed into its recess without misalignment. Should the feeding hole be displaced from the central axis of the riser, the heat axis will move away and the steel will solidify in the riser earlier than in the casting and the riser will fail to fulfil its purpose.

25. DRYING THE MOULDS AND CORES USED FOR STEEL CASTINGS

The moulds and cores used for steel castings are dried to promote the strength of the walls made of sand, clay or other similar mixtures and expel the surplus moisture which may stimulate the formation of gas by foundry moulds. The mould or the core can be dried through, on the surface or else seasoned when they are made of preheated or cold cement mixtures.

The theory of drying and the operation of drying ovens are described in other parts of this textbook. This section will deal only with the specific features of the process observed in moulding steel castings, whose moulds are subjected to a more intensive heating by liquid steel and a higher pressure head, and also to the erosion by a heavy stream of steel passing through the mould channels at a higher linear velocity.

The duration of the mould drying should be proportional to the thickness of the heavy section of the mould or core. Each 25 mm of thickness of the sand material requires a holding of 3-4 hours in an oven at the maximum temperature. With the drier in normal operation, the drying conditions for steel casting moulds are determined from their weight and the overall dimensions of the moulds (Table 30).

Table 30

Drying Conditions for Steel Casting Moulds

(Fantalov)

Drying conditions	Casting weight, tons					
	up to 0.5	0 5-1	1-3	3-5	5-10	10-25
	Mould dimensions, mm					
	500×300 to 1,200× ×400	1,500× ×1,500	2,500× ×2,000	3,000× ×3,000	5,000× ×3,000	6,000× ×5.000
Duration of the entire drying process, hr, including:	6-8	8-12	16-20	18-24	24-36	36-48
placing the mould on the trolley	up to 0.1	0.1-1.0	1.0-1.5	1.5-2.0	1.5-2.0	1.5-2.0
heating	2-3	3-4	5-7	6-9	8-12	10-13
holding	3-4	4-6	9-10	9-11	13-20	20-30
secondary drying	0.5	0.5	1.0-1.5	1.5-2	2-4	6-8
Fuel consumption, kg/m³ for the first drying cycle . . .	5-6	6-8	7-9	8-10	10-12	12-15
for the second drying cycle . .	0.5-1	0.5-1	1.0-2	2-3	3-4	4-5

Chamotte moulds should be dried and calcinated at relatively low temperatures (600-900° C). The secondary drying is done twice at 180-250° C after swabbing the mould. The wash is applied at 60-80° C, since the coating of cold moulds without sub-

sequent drying leads to the formation of blowholes in the castings while the washing of too hot moulds spoils the surface of the castings.

The cores are dried in the usual manner depending on the kind of the binder. Skin drying is used for large castings moulded in floor or in large moulding boxes which are difficult to move into and pull out of the drying oven. Skin drying is performed with gas burners or torches. The moulds are calcined in special heat-treatment furnaces.

The up-to-date process of moulding in sand and clay mixed with soluble glass dispenses with drying ovens and furnaces which encumber the moulding shops and deteriorate the hygienic conditions in the foundries. The use of bentonite also makes drying unnecessary.

When absolutely necessary the moulds can be dried in electric furnaces.

26. STEEL POURING THEORY AND PRACTICE

It has been mentioned in the preceding sections that a solid structure of metal is obtained by raising it to such temperature as would enable it to fill the remotest sections of the mould before solidification sets in. For this, the steel removed farthest from the gates should not attain the liquidus temperature. Excessively superheated steel is liable to cause defective surface of cast metal due to burning-on and cracking. Without dwelling again on the defects which were discussed in detail above, we must point out that each steel and each casting and ingot should have its own maximum range of pouring temperatures (see Table 17).

The pouring process aims at obtaining high quality metal in all the moulds. In the past, only skilled foundrymen with a vast production experience could tackle this complex task. But even they could not obtain good castings in the first and last moulds. The metal tapped from the furnace must have suitable chemical composition with inclusions strictly within the limits established for the given grade, and as little gases and nonmetallic inclusions as possible. But this forms the subject of steel metallurgy which is characterised by its own specific features in the production of shaped steel castings outlined at the end of this textbook. Here, we are interested only in the pouring conditions from the standpoint of thermotechnics.

The ladles intended to hold liquid steel must be thoroughly dried after repair and raised to a sufficiently high temperature (not below 800-900° C and sometimes even higher). The steel tapped from the furnace usually cools in the ladle by 40-100° C prior to pouring. The smaller is the ladle capacity and the lower the initial temperature of the refractory lining, the faster are the

rates of cooling. Insufficient temperature of the ladle lining causes high heat losses and skulling in the ladle involving difficulties in removing frozen metal. If the temperature of steel is too low it may solidify under the stopper and in the nozzle hole and cause various defects and even breakdown when the hole is opened.

The steel should be obviously poured from the ladle at a temperature above the liquidus to provide for a sufficient reserve of heat expended on heating the walls and channels of all the moulds from the start to the end of the pouring process.

Several theoretical principles of liquid metal pouring can be outlined.

1. The heat content of the molten metal tapped from the furnace should correspond to the heat given off in pouring into the ladle, heating the ladle refractory lining to the temperature of liquid steel, pouring the metal into the moulds and heating of their walls.

2. Since some heat is lost during pouring the heat content of the metal should be so high that its temperature in the remotest parts of the casting during the filling of the last mould remains equal to the liquidus temperature or is slightly above.

3. The heat input (heat content of the metal in tapping into the ladle) should be balanced by the heat lost by radiation in pouring, and in heating the ladle and the moulds being filled. This can be expressed by the heat balance equation

$$Q \approx Q_1 + Q_2 + Q_3 + Q_4, \qquad (190)$$

where Q = relative heat content of superheated metal before tapping;

Q_1 = radiation losses during tapping of metal from the furnace into the ladle;

Q_2 = heat lost by radiation in the ladle and in heating the ladle lining;

Q_3 = heat lost by radiation in pouring the metal into the moulds;

Q_4 = heat lost during the motion of liquid metal in the moulds.

The heat content of metal before tapping from the furnace can be found from the equation

$$Q \geqslant G_1 \gamma_1 [c_1 (\theta_i - \theta_l) + \varrho], \qquad (191)$$

where G_1 = weight of metal during tapping into the ladle, kg;

γ_1 = specific weight of liquid metal, kg/cu m;

c_1 = heat capacity of liquid metal, kcal/kg deg;

θ_i = initial temperature of metal during tapping from the furnace into the ladle, °C;

θ_l = liquidus temperature, °C;

ϱ = latent heat of solidification, kcal/kg.

The terms in the right side of the equation can be calculated on the basis of the well-known laws of thermotechnics which were utilised to calculate the temperature, time and rate of cooling of the metal. These calculations fail to provide, however, conclusions as to the quantitative advantages. Eq. (190) characterises the metal balance only qualitatively, generalising all the phenomena which attend the existing pouring processes. A low superheat of steel before the whole of steel is tapped from the furnace should be avoided since the last moulds will be filled at an imminent hazard of developing short runs, oxide films, high residual stresses, blowholes, etc.

Excessive superheat will entail other harmful consequences mentioned above and will require extra heat energy and time for a heat thereby deteriorating the economic indices.

The heat losses expressed by the terms in the right side of Eq. (190) are proportional to the surfaces of the metal being poured. The heavier are the castings the lower should be the metal superheat described by the left side of the equation. Several conclusions can be drawn from this.

1. The length of the stream of steel flowing along the spout into the ladle should be as short as possible, in other words, the ladle should be placed as near to the end of the spout as practicable. The stream should be uniform, without sputtering and bifurcations and directed at the bottom of the ladle far from the stopper rod but not at the ladle wall. In the same manner, when pouring steel from the ladle the stream should be directed at the hole in the pouring cup by the light spot bringing the ladle as close to the mould as possible.

2. Before pouring, the ladle should be thoroughly dried and calcined, with the temperature of the refractory lining brought to its maximum. It is good practice to raise the ladle lining to red and sometimes to yellow heat.

3. Since the heat losses are especially pronounced in small-capacity ladles they should be so designed as to reduce the heat losses by radiation to the maximum. Drum ladles preserve the heat of liquid metal better than open-type ladles.

4. In large ladles the thick layer of slag reliably protects the metal from heat loss by radiation. For this reason, no slag should be detained in the furnace during metal pouring. The layer of the slag above the metal should be not less than 100 mm thick.

5. The highest heat loss is described by the term Q_4 in Eq. (190), i. e., the loss due to heating the channels and walls of the mould. It follows from this that when the metal is poured into moulds for thin-walled and large-size castings the initial heat content of the metal should be higher than in the case of heavy castings.

6. In filling dry moulds less heat is lost than in filling identical green-sand moulds and the heat content of steel in the latter case should be higher than with dry-sand moulds.

7. Moulds surfaced with low heat conductive coatings made of silica flour, etc., preserve the heat during pouring much better. The heat conductivity of the backing sand should be higher for heavier castings to cut down the cooling time of the casting in the mould.

8. Correct determination of the temperature of the molten metal before tapping and during pouring is the main prerequisite for a proper control of the thermal processes.

Optical pyrometers widely used in foundries to measure steel temperature give an accuracy of $\pm 40°C$. The accuracy of the optical pyrometer somewhat improves if the temperature of the metal in the furnace is measured by means of an open end of a pipe immersed into the bath. The measurement must be done very quickly before the pipe is bent or fused.

Ukrainian foundrymen place a cold rod into the bath to find whether the steel is ready to be deoxidised. If the metal is heated well it will quickly begin to boil around the rod. The intensity with which carbon oxide bubbles evolve shows whether deoxidisers may be introduced into the metal. The foundrymen in the Urals take a metal sample with a spoon and pour it down onto a plate from a height of 500-600 mm. A well heated metal produces a thin uniform stream and rebounces in fine sprays while a cold metal has an interrupted stream and large spattering drops.

K. Voronov and A. Silayev in cooperation with the foundrymen at the Novo-Sormovo Engineering Plant developed a highly efficient method to determine the readiness of electrical steel for tapping by a fluidity test. Steel was sampled from the furnace bath with a spoon with an aluminium wire inserted into it. Then the metal was poured into the mould with Ruff's straight-bar test piece.

Excessive superheat of steel in the furnace is not only useless but in a large measure harmful since a strongly superheated steel contains a large proportion of dissolved gases. Superheat does not affect in any material way the mechanical properties which may be corrected to a certain extent by appropriate heat treatment.

Before tapping the steel temperature reaches its maximum. After tapping is completed the temperature as measured in the ladle decreases gradually by 30-100°. The rate of this decrease depends on the temperature of the ladle refractory lining, its capacity and shape. In ordinary open-type ladles when the metal is covered by a layer of slag its temperature decreases at the following average rates:

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Ladle capacity, tons	Freezing rate, °C/min
0.18	20-60
0.70	15-20
·5	15-12
10	4-8
30	3-7
60	2-4
100	1-2

Holding of steel in the ladle before filling large moulds will do much good in the way of reducing gas saturation several times. The holding time depends on many factors, such as grade of steel, number of moulds and ladle operations, thickness of the mould walls and pouring temperature. Holding in the ladle can be recommended for some grades of stainless and heat-resistant steels, in pouring wheel-type castings tending to develop hot cracking, and in pouring heavy castings without internal chills. Too long a holding may freeze the nozzle hole which will have to be pierced with an oxygen torch and the resulting frozen metal in the ladle will have to be removed by the casters.

The ladle ready to receive metal should be free from the frozen metal stuck during the previous operation. New frozen metal is apt to develop quickly in a cold ladle, especially in haste, when the preparation of the ladle has been delayed and the ladle is brought to the furnace without preheating. The frozen metal will stick to the stopper and the stopper rod will not separate from the nozzle. Piercing the hole with an electrode or oxygen torch, or placing it on a stake, constitute a case of emergency.

At first, steel should be poured into large moulds to heat up the cold solid layer of steel in the bottom of the ladle. Small moulds are ordinarily filled at the end of the pouring operation because less metal is lost in this case and the hazard of short run in the last mould is less. The measurement of the metal temperature in large ladles with immersion thermocouples shows that the steel seemingly heats up towards the end of pouring. This is not actually the case. The fact of the matter is that the cold particles of metal sink down due to convection and the warmest particles rise to the surface.

Apart from controlling the heat content and temperature of the steel, no less attention should be given to the other similarly vital factor in the pouring process—the pouring rate.

In this respect, tipping ladles for small and large-size but thin-walled castings are unparalleled. By tipping the ladle the pouring rate can be easily adjusted as desired. Such ladles are used predominantly to pour acid steel, the viscous slags of which are easily retained at the ladle lip. The basic slags remain

fluid almost during the entire pouring process and it is a much more difficult matter to retain them in the ladle.

The rate of pouring from a stopper ladle may vary only within very narrow limits. At the start of pouring the level of steel and the pressure head are higher than at the end. A second stopper device may be provided to double the pouring rate if necessary.

The moulds prepared for filling are grouped according to the steel grades and arranged in one line by the centres of the pouring cups. First the larger moulds should be filled and then the smaller. This arrangement of the moulds is extremely important as the pouring is done at the minimum loss of time on the ladle travel and the metal loses less heat by radiation.
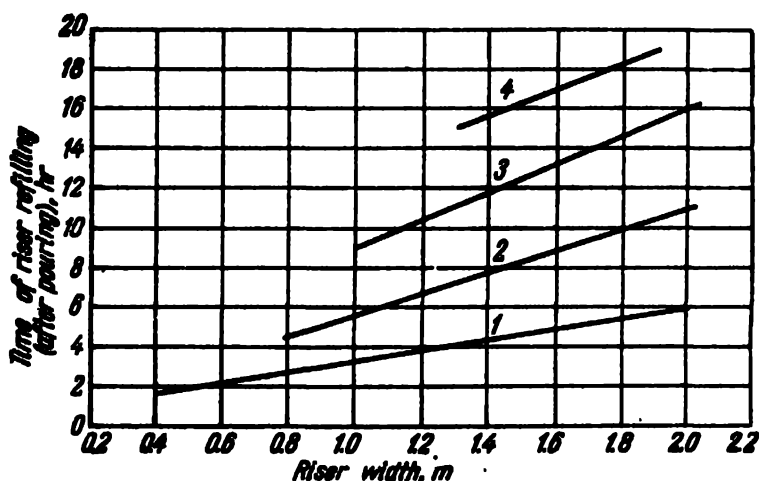
Fig. 119. Time schedule for refilling the risers for heavy steel castings (Vasilyev):
1-4—ordinal number of refilling operation

Immediately after pouring the projection cods, which hamper contraction, are loosened in closed moulds, the gates are removed as well as the risers while they are not yet dark and can be easily separated. In some of the moulds the risers should be refilled. This is done when and as required, but it is as important an operation as a properly organised pouring process.

The risers are refilled as shown on the diagram in Fig. 119. The Y-axis is used to plot the time after the end of pouring and the X-axis—the riser width. The first repouring is done on the basis of the weight of the unfilled part of the riser; for the second the weight of the added metal is determined as 60 per cent of the difference between the specific gravities of the solid and liquid metal, taking the entire weight of the casting and the riser, and 40 per cent of this difference is taken for the third repouring. Small risers 400 to 800 mm in width are refilled only once.

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Very large moulds are refilled when a thick layer of metal has solidified on the riser surface. The layer is firstly cut through with an oxygen torch and then the riser is filled to the top from the ladle.

Special measures should be taken to pour steel into small moulds. The heat loss during the pouring process is so high that no superheating of steel in the furnace will be of any avail. An example is provided by a conveyor-type casting. The thickness of the casting wall is 5 mm, an average weight is 0.15 kg and the mould capacity is 5 to 6 kg. The steel is smelted in a converter with a burden weight of 1.5 tons. At the end of blowing, the steel is tapped at 1740°C into a crane ladle and is delivered from it by means of hand ladles of 40 kg capacity. While being repoured into the ladle the temperature drops by 100°C. It has been found unexpedient to compensate for such cooling by a still higher superheating of the steel in the converter since this required extra ferrosilicon and refractories.

The correct solution was found by building an oven fuelled by coal for preheating the ladles. The refractory lining of the ladle was heated to 1150°C. The temperature drop during re-pouring decreased by 40°C as the ladle was filled with the metal for the first time. After that the lining was heated with liquid metal to 1300-1400°C and the subsequent drop in the temperature did not exceed 20°C. The hand ladles were replaced by a tea-pot spout ladle designed without a stopper. This greatly facilitated the work of the casters and increased the output by 20 per cent.

Obviously, a drum ladle will preserve heat under these conditions more effectively while in a tea-pot spout ladle some heat is lost for heating the partitioning wall.

27. COOLING OF CASTINGS. COOLING RATES. CALCULATION

When a casting is cooled in the mould not uniformly the temperature is different in various parts of its volume. The temperature difference and the thermal stresses increase in proportion to the difference in the thickness of the walls of the casting (see p. 119). The square root equation (p. 59) shows that a casting with a thin average, or referred, wall thickness will cool at a faster rate and can be knocked out of the mould by a shake-out grid at an earlier date. A thick casting takes a longer time to cool and the knocking-out operation is delayed.

Let us illustrate our example by the cooling of a casting of a steel die bed weighing 45 tons and 1,560 mm thick. The temperature measurements taken at various depths from the surface yielded isotherms plotted after 1, 9, 14 and 20 days and isochores for 1000°C (Fig. 120). At the moment of the predominant development of elastic deformations in the central zone of the bed the

temperature difference in the internal and external zones amounts to 180°C. If after this moment the casting is allowed to cool in the mould for 5 to 6 days more the temperature difference and the stresses will remain the same even if the external zone reaches a temperature of 100°C.

Some heavy engineering plants use a rate of 3 tons per day for cooling large castings in the mould. This clearly exaggerated and ungrounded rate reduces the operating capacities of the moulding shop.
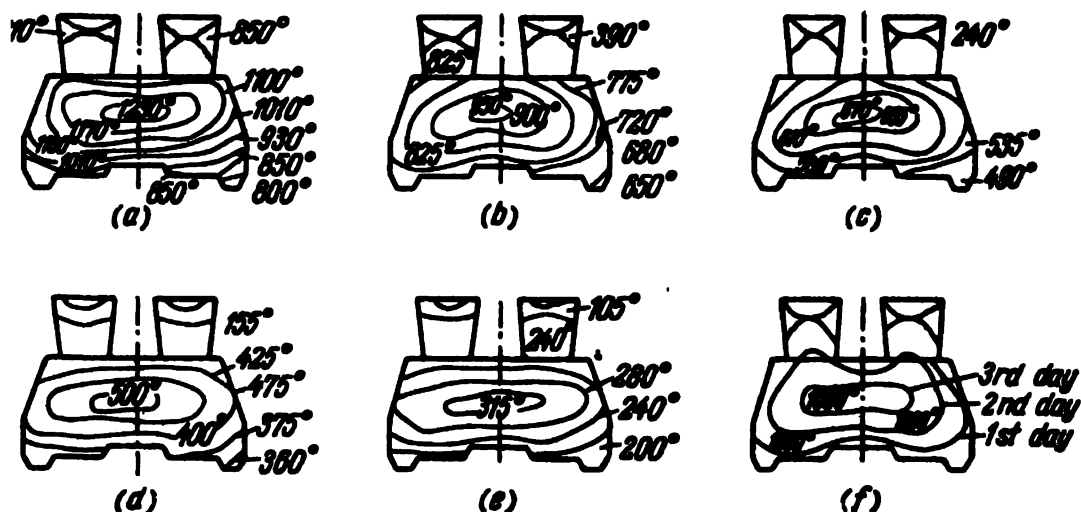
Fig. 120. Isotherms in various zones of the volume of a casting for a steel die bed weighing 45 tons at different time after filling the mould (Baranov, Lipov):

a—isotherms of the first day; b—isotherms of the fourth day; c—isotherms of the ninth day; d—isotherms of the fourteenth day; e—isotherms of the twentieth day; f—isochores for 1000° C

The heat transfer in a sand-clay mould can be expressed as a ratio of the coefficients of temperature conductivity, or heat accumulation, δ, for the casting b_1 and the mould b_2:

$$\delta = \frac{b_1}{b_2} = \frac{\sqrt{\lambda_1 c_1 \gamma_1}}{\sqrt{\lambda_2 c_2 \gamma_2}}, \qquad (192)$$

where λ_1 = heat conductivity of the metal amounting to 25 kcal/m² hr per deg, as shown by laboratory tests;

c_1 = heat capacity of steel equal to 0.2 kcal/kg deg;

γ_1 = specific weight of steel equal to 7,600 kg/m³;

λ_2 = heat conductivity of a sand mould equal to 0.64 kcal/m² hr per deg;

γ_2 = specific weight of the mould equal to 1,600 kg/m³;

c_2 = heat capacity of the mould equal to 0.2 kcal/kg deg.

The accumulation coefficient is $b_1 = 170$ for the steel and $b_2 = 17.0$ for the mould. The temperature conductivity δ is 0.013 hr/m².

The results of investigating fifty castings weighing from 60 kg to 130 tons cooled in sand moulds were used to compile diagrams for calculating the cooling time which showed when the castings must be withdrawn from the mould. The castings of any shape are placed within three classes, accepted in thermotechnical calculations, namely: class I includes plates, class II cylinders or bars, and class III cubes and balls.

Table 31 subdivides all castings into classes and groups depending on their characteristics and purpose. The recommended temperature of withdrawal from the mould after pouring and freezing is indicated in the last column of the table, in conformity with the current practice at heavy engineering plants. The temperature of withdrawal has been determined very carefully to prevent the formation of cold cracking. The safe temperature for the extraction of uncritical castings can be set at 900°C (light-red heat). This category includes pile driver rams, flooring plates, weights for moulding boxes, etc. On the other hand, thin-walled parts with abrupt changes from thin to thick sections should be withdrawn at the minimum possible temperature, not above 100°C. Here belong the rims of toothed gears, wheel-type parts, ribbed panels, etc.

The time for the withdrawal of the castings after the mould had been filled and the castings solidified was determined by special temperature measurements to find the temperature ranges under the usual practice of casting in dry and green sand moulds.

The calculations were used to draw a simplified diagram (Fig. 121) which showed the duration of the cooling period of castings irrespective of the properties and the condition of the sand-clay mixture. The heat conductivity was assumed constant and equal to $\lambda=0.65$ kcal/m hr per deg, the heat capacity $c=$ $=0.22$ kcal/deg and the specific weight $\gamma=1,850$ kg/m^3. For the grade 35Л steel the pouring temperature θ_p was 1550°C.

The weight of the casting with the risers (mould capacity) was plotted on the X-axis. The typical size of the casting (volume-to-surface ratio as determined by the formulas depending on the class of casting) is plotted on the Y-axis. The lines sloping rightwards correspond to the changes in the typical·dimensions of the castings of various pouring classes depending on their weight.

The lines sloping leftwards correspond to the duration of cooling of castings with the given typical dimension to the temperature indicated for the three classes at the top.

Let us consider some examples of calculating the freezing time from the moment the mould is filled and up to the withdrawal of a solid casting.

Example 1. Determine the minimum duration of cooling of the casting for an electric locomotive frame with the weight

of about 8 tons (including the risers) and a thickness of 25 mm and with the mould capacity-casting volume ratio of $\dfrac{V_{cap}}{V_{overall}}=$ =0.28 by overall dimensions. According to Table 31, the casting belongs to class I and ranks among the critical parts operating at a variable impact load. The temperature of withdrawal is 300°C. The castings tend to warp and develop cold cracking.
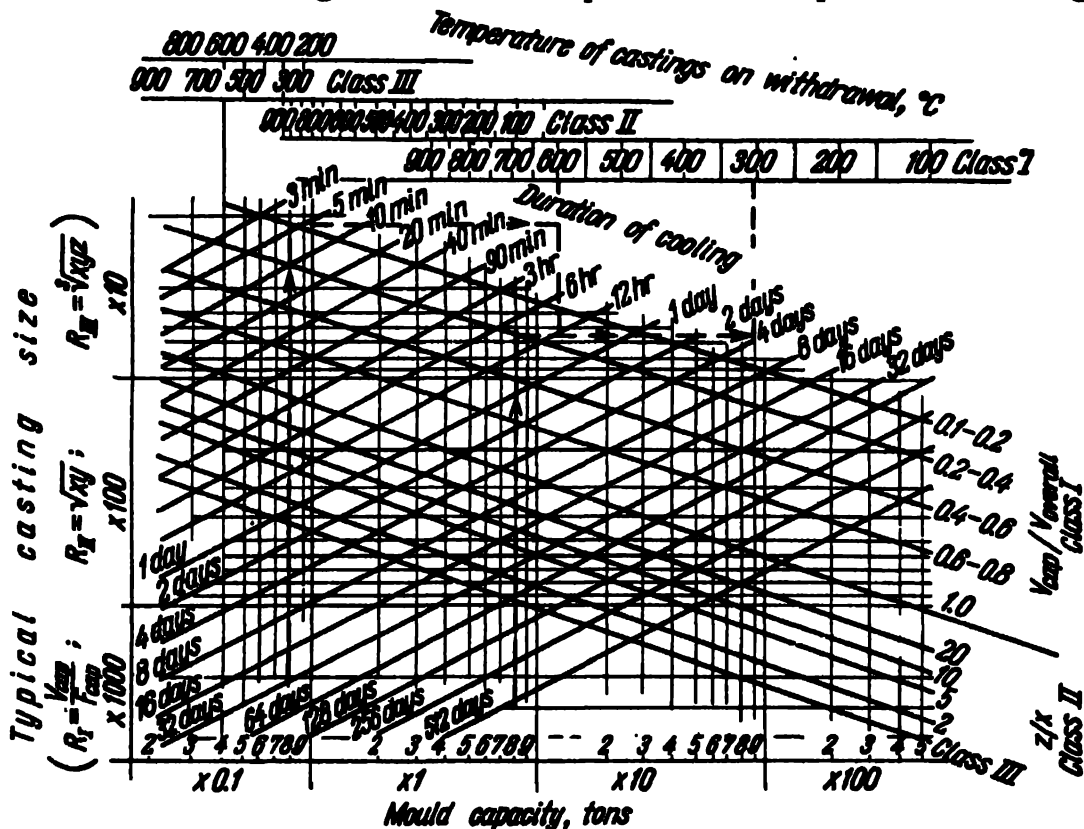
Fig. 121. Diagram for determining the duration of cooling of steel castings to the temperature of withdrawal (Novikov, Gruzin)

Let us erect a perpendicular from point *8* in the middle group of straight lines to the intersection with the line sloping rightwards which corresponds to the ratio $\dfrac{V_{cap}}{V_{overall}}=$0.2 to 0.4. Let us draw a horizontal line from the point of intersection to the right to the vertical dropped from point *300* on the lower of the top scales belonging to class I castings. The duration of cooling to 300°C will be about four days.

The withdrawal at 600°C will require a cooling period of only 12 hours.

Example 2. Find the temperature of a gear with the risers weighing 800 kg and 60 mm thick if it is withdrawn 90 min after pouring. The casting belongs to class I. The ratio $\dfrac{V_{cap}}{V_{overall}}=$0.1. Let us draw a perpendicular to intersect the line 0.1-0.2. Then, we continue the line of 90 min to the horizontal plotted from the

Table 31

Temperature Recommended for the Withdrawal of Castings Made of Carbon Low-alloy Steel
(Gruzin, Novikov and Shiryaev)

Characteristics and purpose	Castings			Recommended temperature of withdrawal, °C
	Class I (x=y=z)	Class II (x=y≪z)	Class III (x=y=z)	
Parts of simple shape weighing up to 3 tons	Armour plates of coal crushers; flooring plates; housings and caps of bearings; stone crusher jaws	Weights and clamps for moulding boxes; smooth shafts; girders	Pile driver rams, balls of coal crushers	900
Parts with uneven wall thickness and heat centres, cast with external and internal chills	Parts for housings of steam turbines and reducing gears; blades, stators and rotors of hydraulic turbines; cross-pieces and architraves of presses; frames of electric locomotives, cars and earth diggers; hollow rollers of roll tables and pipes; moulding boxes; pouring basins; drums; supports; sheaves and spokeless rollers	Beds for rolling mills; crank-shafts, columns and tierods of presses; rolling and mill rolls	Die beds, roll cushions	600

Characteristics and purpose	Castings			Recommended temperature of withdrawal, °C
Critical parts operating at variable impact loads tending to warp and develop cold cracking	Cylinders of piston engines and presses; plough shares; marine propellers	Gears, flywheels, sheaves with spokes, toothed racks, cast propeller shafts	Hammer heads, dies	300
Thin-walled parts with abrupt changes to heat centres (cast tools)	Fancy steel casting, side milling cutters	Toothed rims, plain and helical milling cutters, screw drills		100

point of intersection of the preceding straight lines to the right. From the new intersection point we draw a vertical to intersect with the bottom temperature scale of class I castings. The temperature of the gear will be 600°C.

The temperature at which the castings are withdrawn can be specified more accurately from other temperature measurements and from the observed quality of the castings. The following factors affect the temperature of withdrawal:

(1) the greater the nonuniformity of cooling of class I castings, the lower should be the temperature of withdrawal;

(2) higher temperature of withdrawal can be taken for simple castings which do not warp;

(3) higher temperature of withdrawal can be also permitted in conjunction with special measures intended to eliminate restrained shrinkage, for example, when solid cores are replaced by hollow cores.

During the two-thirds of the time of full holding prior to knocking out, the mould is left at rest except for one operation of breaking off the risers while they are still hot. This operation should be well organised so as not to spoil the entire casting. The temperature of the necked-down risers with wafer cores should be in the vicinity of 1100-1200°C. The risers are much more difficult to separate at red heat.

As has been pointed out the moulding boxes are removed from the mould approximately after two-thirds of the time elapsed from pouring while during one-third of the time the casting remains in the burned-on sand.

The removal of hot mixture from the casting is a difficult operation injurious to the health of the knockers-out, and every possible measure should be taken to make the labour conditions healthy.

1. The knocking-out operation should be mechanised to the largest extent possible and done on jolting knock-out grids.

2. The hot gases and abundant dust should be disposed of by good ventilation. It is frequent practice to provide strong streams of air sucked in through the slots in the jolting grid. Heavy castings also require wetting with water by sprayers. When water gets onto a casting cooled to 200-300°C no additional stresses will be set up in the metal. The wetting and evacuation of gases and vapours considerably improve labour conditions and increase labour productivity.

3. At some plants in Czechoslovakia the hot mixture is removed from the jolting grids and delivered to the reclaiming system pneumatically along water-cooled pipes. The heat is utilised in the central heating system.

4. The cooled mixture is sifted and all lifting hooks and fragments of metal are separated with magnets and delivered into

the hopper for used mixture which goes to make backing and facing sands.

5. Pneumatic pipe conduits should be provided with a device for cleaning off the sticking mixture.

6. When the ventilation facilities are too weak the knocking-out section should be housed in separate premises to keep the dust off other shops.

28. CLEANING. REMOVAL OF GATES AND RISERS

The cleaning of castings knocked out from the moulds is an extremely difficult operation which requires adequate measures to be taken to protect the personnel against abundant silicate dust.

The greatest effect in cleaning steel castings from burnt-on mixture is derived from hydroblasting in special chambers by means of adjustable nozzles hinged on ball joints which direct a jet of water mixed with sand onto the surfaces of the casting at a pressure of 100-120 atm.

At the Uralmash Plant the hydraulic chamber is connected with a device for a wet reclamation of moulding mixtures and has a very compact design. A truck with a carrying capacity of 50 tons is pushed into the chamber. Larger castings are delivered through the detachable upper roof. The diameter of the nozzle outlet is 5-6 mm. The nozzle is positioned 100-150 mm away from the casting. The truck can swivel vertically through 360°. The washed-off pulp passes through a sizing separator. The gates consumption is 30 m³/hr. The productivity of the reclamation unit is 4 m³/hr, including 1.25 m³/hr from the hydraulic chamber and 2.75 m³/hr from the hopper with used mixture. Some 15 m³/hr of water is needed to wash off the mixture from the castings.

Shot blasting cleaning chambers used in small-lot production are markedly inferior to the above design as far as hygienic conditions and efficiency of the process are concerned.

After all sand has been removed from the castings they are finished with abrasive disks. The gates are cut off by diverse method. Small castings are processed on eccentric presses. In serial production use is made of turret semiautomatic machines (up to 100 clamping fixtures) which enable about 2,000 small different castings to be processed during one hour.

The gates are separated from large castings by pneumatic cutting hammers with subsequent finishing by abrasive disks with a flexible shaft. Sometimes the gates are cut off by type УР oxygen torches. Type УР hand cutters are employed to cut off the risers. But this method gives a poor surface and requires large machining allowances.

A recent development is a universal portable tip for the ОП-1 cutter for removing the risers from steel castings 100 to 500 mm in diameter, which eliminates the shortcomings inherent in hand cutting. The tip can be secured on the prop of a truck, on the jib of a crane, on a cantilever and other supports.

The consumption of oxygen varies from 20 to 60 m³/hr depending on the diameter of the riser to be cut off while that of acetylene between 1,100 and 1,750 lit/hr.

A gas torch can be effectively used to clean castings instead of abrasive disks, saving labour and improving production conditions.

29. HEAT TREATMENT OF CASTINGS

The heat treatment of steel castings is characterised by the initial and final temperatures, the rate of the temperature variations, the time of holding the product at the given temperature, the initial and final structure and the properties of steel castings.

According to Soviet standards, the castings of quality class I can be shipped, on special agreement with the user, untreated. Higher class castings are subjected to various heat treatment operations.

The heat treatment of castings made of carbon or alloy structural steel depends on the purpose they are intended to serve. Most widespread is the so-called full annealing to improve the mechanical properties of the cast metal without further hot mechanical processing. Homogenising is resorted to when the ductile properties of steel should be improved.

Sometimes annealing is called upon to decrease the hardness of the castings and improve their machinability. This process is known as softening. The internal stresses set up in shaped castings are relieved by process annealing.

To improve its mechanical and especially ductile properties, the steel is annealed at a temperature above the critical range of structural transformation.

Very often, each of the above kinds of annealing may have a nature of preliminary processing followed by finish treatment — normalising, quenching and tempering and high tempering, and ageing and structure stabilisation for certain grades of austenitic steel. The finish heat treatment is employed after the risers have been cut off with torches. Given the consent of the user, carbon steel of grades 15Л, 20Л and 25Л may be delivered untreated after cutting off.

Thin-walled simple castings may be finish heat treated without preliminary processing.

The castings are heat treated in special car-bottom furnaces, placed on the bottom so as to leave sufficient space for furnace

gases. Shaped castings may buckle or sag owing to their weight and should therefore be supported by firebrick piles. Castings made of carbon or alloy steel without heavy cross sections may be charged into a furnace preheated to the necessary temperature and heated intensively.

After the temperature has been equalised in the central portion of the burden the castings are held at the given annealing temperature at a rate of 1 hr per each 25 mm of the thickest section.

After holding, the castings are cooled with the furnace down to 300-400°C and pushed out into open air.

Normalising is used as a combined operation of preliminary and finish heat treatment. After holding the castings for normalising they are rolled on the car bottom out of the furnace and cooled in air.

The castings made of alloy steel, except for grade Г13Л, are quenched by immersing them in a tank with boiled oil, and simple castings of carbon steel and of steel Г13Л are placed into a tank with running water.

To prevent the formation of annealing cracks after quenching the castings should be slowly transferred into a heated furnace for tempering.

Table 32

Holding Temperatures in Heat Treating Castings Made of Carbon and Alloy Steels

(Gulyaev)

Grade of steel	Temperature, °C				
	Annealing or normalising in preliminary heat treatment	Normalising		Quenching in oil	
		Normalising proper	Tempering	Quenching proper	Tempering
35 Л	—	860-880	600-650	860-880	600-650
40 Л	—	860-880	600-650	860-880	600-650
35 ГЛ	—	—	—	850-870	600-650
40 ХЛ	—	—	—	850-870	600-650
40 ХНЛ	—	—	—	850-870	600-650
35 ХГСЛ	—	870-890	550-600	—	—
35 ХМЛ	880-900	860-880	600-650	860-880	600-650
30 ХНМЛ	880-900	860-880	600-650	860-880	600-650
30 ХНВЛ	880-900	850-870	600-650	860-880	600-650
30 ДХНСЛ	880-900	860-880	600-650	860-880	600-650
40 ХНТЛ	—	—	—	830-850	550-600

The duration of holding at the tempering temperature for castings 25 mm thick is 2 hours, while 30 min should be added besides for each 25 mm of thickness in the case of heavy castings. After tempering the castings are cooled in air, except in the presence of possible temper brittleness when after holding for tempering they should be quenched in water.

Table 32 recommends holding temperatures in heat treating steels of certain grades.

Fig. 122. Change in the mechanical properties of castings made of grade 36Л steel depending on heat treatment procedure and the thickness of the casting wall

The typical mechanical properties of cast carbon steel of grade 35Л and certain grades of alloy steel after normalising and quenching, obtained by B. Gulyaev on the samples cut out of the test bars with various wall thicknesses, are illustrated in Figs 122, 123 and 124. The test bars were quenched or normalised at 850-880°C, with subsequent tempering. After various degrees of tempering the mechanical properties were assessed from 30-mm bars and those from the bars of various thickness after tempering at 650-670°C.

The illustrated diagrams show that a higher content of alloying elements in steel considerably increases the yield point and ultimate strength at the constant ductility—elongation and reduction of area. The impact strength increases but slightly. The yield point-to-ultimate strength ratio increases too.

As compared to normalising, quenching tends to increase the yield point and the ultimate strength at the same elongation and reduction of area. This difference grows smaller, however, with a higher degree of alloying.

Shaped castings made of alloy steel should be normalised and tempered to reduce internal heat stresses and not quenched and tempered.
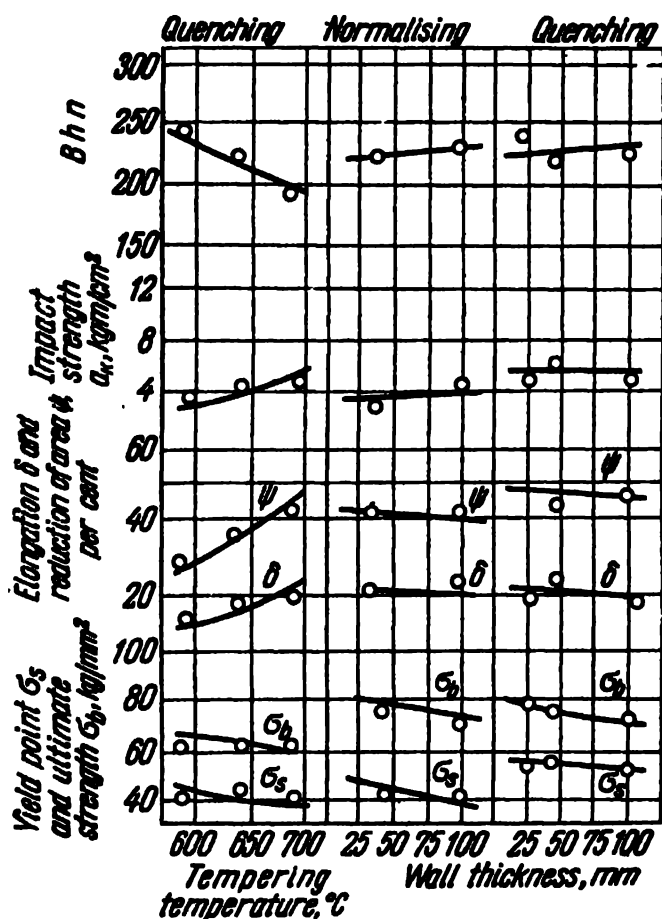

Fig. 123. Change in the mechanical properties of castings made of grade 40ХЛ steel depending on heat treatment procedure and the thickness of the casting wall (Gulyaev)

An increase in the thickness of the wall of the casting or the test bar from 12 to 100 mm brings about a certain reduction in the mechanical properties, except for impact strength which remains the same. The elongation decreases by 15-18 per cent and the reduction of area by 25-30 per cent.

The course of physical metallurgy and heat treatment of steel tells us that when the cooling rate is increased within the critical temperature range of transformations $\gamma \rightarrow \alpha$ the primary grains of cast steel become refined in proportion to this rate. This has been proved by the work of Hanemann and Schrader.

The refining of grain is necessary to increase the yield point and ultimate strength, which, in the final analysis, makes it possible to make the weight of the structure lighter.

According to Soviet standards for castings made of alloy structural steel (19 grades), σ_s ranges from 25 to 60 kg/mm²

after normalising and tempering, and from 35 to 70 kg/mm² after quenching in a liquid medium, with an impact strength of 4 kgm/cm².

Hence the following important practical conclusion: although finish heat treatment of castings in the steel foundry requires extra production space and double the number of the furnaces as well as the space for installing quenching tanks, it gives noticeable advantages due to the lighter weight of the castings.
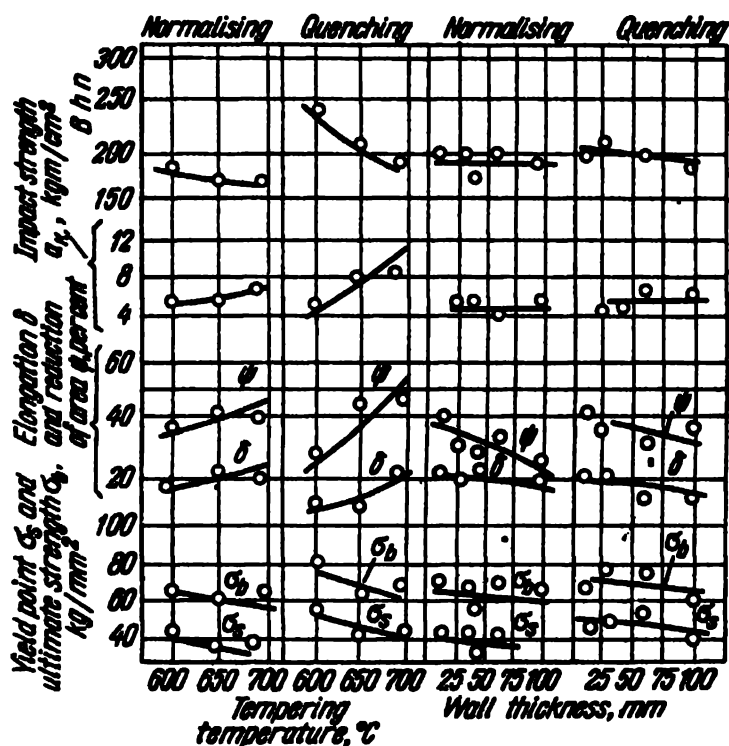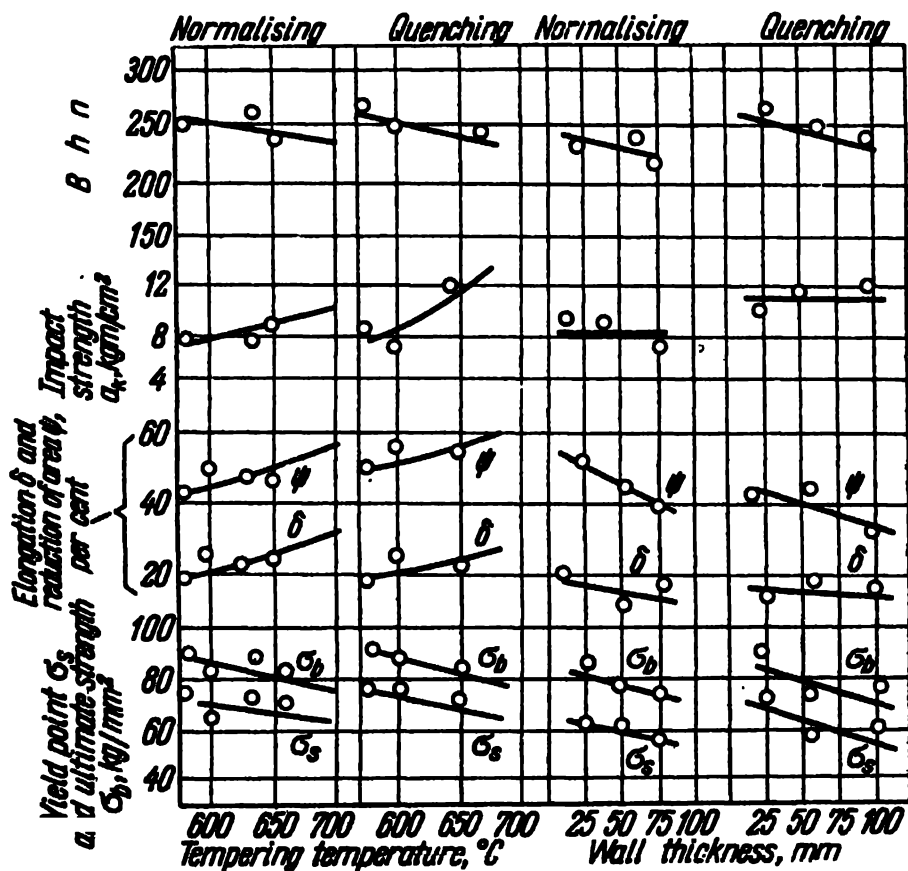

Fig. 124. Change in the mechanical properties of castings made of grade 30ХНВЛ steel depending on heat treatment procedure and the thickness of the casting wall (Gulyaev)

Y. Nekhendzi has shown that when the wall thickness is increased from 200 to 400 mm even in castings made of grade 40ХНТЛ steel, the yield point of the normalised castings drops by 30 per cent and of the castings quenched in oil by only 15 per cent. The elongation and the impact strength decrease twofold in this case.

The wear resistance of castings intended, for example, for the gears of coal crushers made of steel grade 30ХГСЛ (30 per cent C, 1 per cent Mn, 0.6 per cent Si and 0.7 per cent Cr) is improved by normalising from 880-900°C and tempering from

400-430°C. The gears obtained weigh 900 kg at $\sigma_s \geqslant 35$ kg/mm^2, $\delta_5 \geqslant 14$ per cent and $\psi \geqslant 30$ per cent. According to Y. Nekhendzi, high mechanical properties can be obtained in castings made of grade ХГСЛ steel (30 per cent C, 12 per cent Mn, 1 per cent Si, 0.3 per cent Cr, 0.5 per cent Ni and 0.03-0.07 per cent Ti) if normalising is replaced by quenching in water from 830°C and tempering from 650°C in air, i. e., $\sigma_s > 50$ kg/mm^2, $\sigma_b > 73$ kg/mm^2, $\delta_5 > 10$ per cent, $\psi > 20$ per cent and impact strength of over 5 kgm/cm^2. This heat treatment procedure can be applied to small and rather simple castings, such as brackets, rolls and disks. The castings with a high resistance to wear require complex alloy steel of grade КДЛВТ, normalised from 930°C, quenched in water from 930°C and tempered from 270°C; the impact strength is 5-8 kg/mm^2 and Bhn is 390-470.

Normalising or quenching and tempering is employed for castings made of structural alloy steels subjected to impact loads at low minus and also elevated temperatures.

Castings made of heat-resistant austenitic steel require other heat treatment procedures. For example, steel of grade ЛА-1 is heated to 1100° and held at this temperature to enable the carbides to dissolve, after which its structure is stabilised at 900 and 700°C and, finally, aged at a temperature exceeding the prospective operating temperature by 50-100°C.

Grade Г13Л austenitic structural steel is heated to 1100-1150°C and quenched in water for, otherwise, the carbides on the boundaries of the primary austenite grains will be the focal points of stress concentration, like graphite lamellae in pig iron and will make the part brittle and liable to fail in operation.

30. FINISHING AND INSPECTION

Finishing and inspection are the last operations of the casting process requiring high skill of the workers.

The finishing operations include surface cleaning after heat treatment to give the casting its outward appearance, welding of repairable defects, rough machining and straightening of warped castings.

During annealing and normalising the castings are covered with a layer of scale which can be removed by various methods. Small castings are placed in revolving drums while larger castings are cleaned in shot-blasting chambers. In some cases, very heavy castings can be cleaned by oxygen torches.

Many surface defects appear on the surface after finishing or roughing operations. Small cavities and slag inclusions may be welded in. The defective places are cut or burned out with oxygen torches and then welded in and ground with abrasive disks.

The scale on critical steel castings, such as cylinders of steam turbines, is removed by grinding and polishing. The surface is then etched and the density of the casting is checked by gamma-raying. Internal defects are opened, dressed, welded in, cleaned again and subjected to low-temperature annealing to relieve internal stresses.

Gamma-raying is extremely helpful in revealing all kinds of hidden nonhomogeneity and is broadly used today at many engineering plants. Good results in radiographic examination of steel articles 10-50 mm thick are obtained by the use of the radioactive isotope of cesium 137 and, better still, of iridium 192 and europium 154. When the castings are less than 10 mm thick, low-energy radiation isotopes below 02 Mev can be recommended. For castings 50 to 200 mm thick use is made of cobalt 60 which has a higher radiation energy.
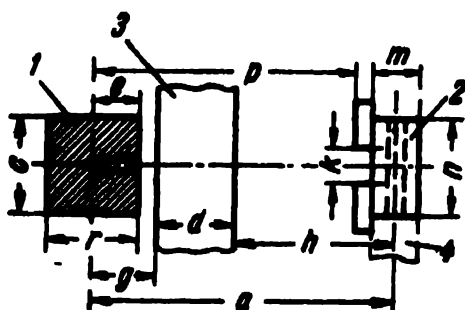
Fig. 125. Design of a gamma flaw detector (diagram)

The sensitivity of a gamma flaw detector used for steel castings 170-200 mm thick is only 2-4 mm, in other words, it cannot be used to reveal defects less than 2-4 in cross section. When the thickness is increased to 300 mm the sensitivity is below 3 per cent. At thicknesses over 300 mm the sensitivity of the gamma flaw detector sharply drops and it fails to reveal defects less than 15-20 per cent.

One design of a gamma flaw detector is shown in Fig. 125. Casting *3* is placed at a distance *g* from radiation source *1* arranged in a container. A narrow beam of gamma-rays pierces the casting, passes through opening *K* onto the screen of counter *2* through amplifier *4* and leaves its imprint on a photographic plate. The flaw detector should move along the surface being examined at a linear velocity not above 4-6 cm/min. The travelling speed of the gamma flaw detector can be increased when the radiation is measured with quick-response type БП-102 potentiometer.

The methods for inspecting the quality of steel castings for mechanical, physical and chemical properties are outlined in appropriate state standards and specifications and in no way differ from inspection of the castings made from other alloys.

CHAPTER IV

OPERATING PROPERTIES OF CAST STEEL

Cast steel differs from forged or rolled steel by certain specific properties. Steel castings are usually covered by oxide films which under certain conditions can protect the metal from chemically active environment and destruction. The mechanical properties of cast steel depend on the composition, the heat treatment procedure and the thickness of the casting wall, which can be attributed to the conditions of solidification, contraction and segregation of admixtures in various parts of a shaped casting.

The operating properties of steel castings are indicated in appropriate State standards and industrial specifications compiled on the basis of observation and production experience of many dozens of enterprises.

The Soviet standards and industrial specifications determine the safe limits for the physical and chemical properties of ready steel castings.

31. CARBON STEEL

Castings made of carbon steel find an extensive application in various branches of the industry. Low-carbon steel of grade 15Л is employed for castings which are to operate at impact loads. Castings for frictional operation (gears, rolls and other similar parts) are case-hardened. The content of carbon in steel to be case-hardened should vary within 0.09-0.16 per cent. Heavier castings should contain less carbon. The content of phosphorus may be increased to 0.08 per cent and that of sulphur not above 0.03 per cent because steel easily develops hot cracks.

Low-carbon steel also goes to make electric motors and dynamotors. It is used to cast jackets, poles and armatures. This steel possesses an increased magnetic permeability and is remarkable for its low hysteresis losses. The combined amount of sulphur and phosphorus should not exceed 0.02-0.03 per cent.

Low-carbon steel is also used to make crucibles for melting magnesium in which case up to 0.20 per cent Al is introduced into the steel instead of silicon.

Since the fluidity of mild steel is low, special measures must be taken to fill effectively large foundry moulds. The formation

of hot cracking is avoided by increasing the manganese content and decreasing the content of sulphur. The moulds should be filled at a fast rate.

The fluidity of medium-carbon steel of grades 25Л-35Л is considerably higher and this gained it the reputation of the universal "casting" steel which is used to manufacture a great variety of castings of various weight, size and purpose. Most parts of medium weight are made of steel obtained in electric arc furnaces with an acid lining. American foundrymen recommend that the content of silicon be increased to kill the steel. Deoxidation of steel with aluminium frequently causes intercrystalline brittleness due to the evolution of sulphides and aluminium oxides. Basic open-hearth steel is deoxidised with aluminium, especially when fine primary grains are to be obtained in large castings. The best results are derived from killing the steel with complex deoxidisers, such as silicocalcium, AMS alloy, etc.

The carbon content should be reduced in the steel intended for frost-resistant castings and the castings themselves hardened and tempered to refine the grains and increase the homogeneity of the metal structure.

To improve the strength and wear resistance the carbon content is usually raised still higher (0.45-0.55 per cent). It is good practice to increase the amount of manganese for such castings as gears, sheaves, flywheels, housings of engine cylinders and other heavy parts. On the other hand, the content of sulphur should be decreased to 0.02-0.03 per cent. If the product is to have a higher impact strength the content of phosphorus should be likewise decreased.

Hardening and tempering build up sorbitic structure in this steel (σ_s=45-50 kg/mm^2, δ_s=30-40 per cent). The steel tends to develop cold cracks and for this reason the sum total of carbon and manganese should never exceed 1.25 per cent.

High-carbon steels (1.0-1.3 per cent C, 0.3-0.4 per cent Si, 0.5-0.6 per cent Mn, <0.03 per cent S, ≪0.03 per cent P) are subjected to graphitisation. Free carbon evolves in the form of graphite in the same way as in the castings made of malleable iron. The impact strength of this steel is below 1-2 kgm/cm^2 and elongation is less than 10 per cent. Castings made of graphitised steel have a rather limited application because of high production costs. This steel goes to make the plates of ball mills and stone crushers and other parts operating in conditions of abrasive wear. The frost-resistance of castings made of graphitised steel is rather low.

When this steel is alloyed it can be used to produce finishing rolls which retain their surface appearance for a long time. The castings are first machined and then hardened and tempered.

In some countries such steel is known under the names of graphmo, graphtung, graphal, graph MNS, depending on the alloying element used. High-carbon steel has rather good casting properties. However, it is rather sensitive to temperature variations and tends to develop cold cracking.

32. MANGANESE STEEL

In α-iron manganese forms solid substitutional solutions when present in amounts of up to 0.10 per cent. The surplus manganese is expended to form the Mn_3C carbide in which iron dissolves practically infinitely. The Fe_3C and Mn_3C carbides are mutually soluble. In an annealed state manganese is present in such steel mainly in the carbide phase and passes to the solid solution of α-iron after hardening.

The manganese content in structural steel can be increased to 1.6 per cent, thereby considerably improving the wear resistance of the working surface of the castings. This steel is liable to develop temper brittleness and the castings heated for annealing or normalising should be rapidly cooled. Low hardenability limits the use of manganese steel to castings with wall thickness under 100 mm. When the manganese content is increased from 0 to 2.8 per cent the Brinell hardness of the ferrite increases from 65 to 140. When the manganese content is increased to 4 per cent the tensile strength continuously increases. The elongation of the steel reaches its highest value at 2 per cent Mn, while the impact strength drops sharply between 0.84 and 2 per cent Mn.

Manganese steel can be easily welded by electric or gas welding methods. The corrosion properties of the steel are rather poor.

Manganese steel is widely used to make castings for railway rolling stock, gear wheels of bridge cranes, earth diggers, road machinery, etc.

Soviet-made couplers for railway rolling stock are cast of grade 27ГЛ steel (0.22-0.32 per cent C, 1.5 per cent Mn, 0.25-0.35 per cent Si, <0.055 per cent S and <0.055 per cent P) melted in a basic furnace. For less critical parts of the coupler, acid steel should contain sulphur and phosphorus not in excess of 0.06 per cent of each while for converter steel the content of sulphur and phosphorus should not exceed 0.07 per cent and 0.09 per cent, respectively.

After quenching in water from 900°C and tempering from 640°C, steel of grade 25ГЛ with 0.25 per cent C, 1.43 per cent Mn and 0.55 per cent Si exhibits the following properties: $\sigma_s = 45$ kg/mm², $\sigma_b = 68.0$ kg/mm², $\delta_s = 21.0$ per cent and $\psi = 53.2$ per cent.

Steel of grade 45ГЛ (0.4-0.5 per cent C, 1.2-1.5 per cent Mn, 0.3-0.45 per cent Si, <0.035 per cent S and <0.04 per cent P)

is used to cast driving wheels and gears and also rubbing parts. After normalising from 850°C and tempering from 550-600°C in the furnace the steel displays the following mechanical properties: $\sigma_s=34$ kg/mm², $\sigma_b=67$ kg/mm², $\delta_5=11$ per cent, $\psi==20$ per cent and $Bhn=196$-236. Steel of grade 40Г2Л with a higher manganese content goes to cast parts for road machinery. After annealing from 870-890°C, quenching in oil from 830-850°C and tempering at 350-450°C and cooling in air the mechanical properties of this steel are almost the same as in steel of grade 45ГЛ while the hardness is a little better (190-255).
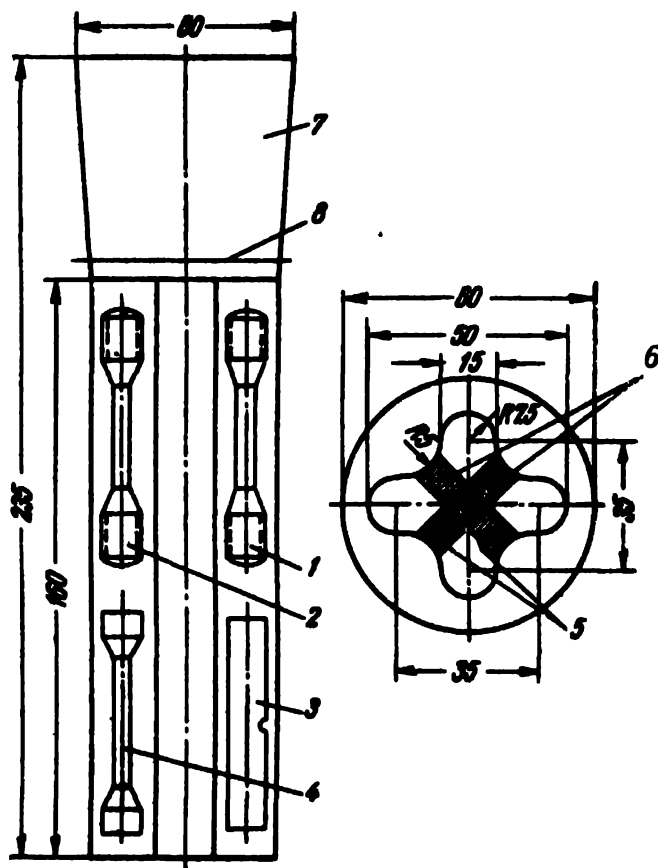
Fig. 126. Ingot for cutting out specimens for testing mechanical properties:

1 and *2*—for a short-term tensile test; *3*—for an impact strength test; *4*—for a prolonged test at high temperatures; *5*—cutting planes; *6*—zone of centreline porosity; *7*—riser; *8*— riser cutting plane

Some plants employ manganese steel of grade Л40Г2 (0.37-0.45 per cent C, 0.2-0.37 per sent Si, 1.4-1.8 per cent Mn, ≤0.3 per cent Cr, ≤0.40 per cent Ni, ≤0.04 per cent S and ≤0.04 per cent P).

The steel is made in electric arc or induction furnaces. The chemical composition may deviate from the specified norms as follows: carbon by ±0.01 per cent; silicon, manganese, chromium, nickel and vanadium by ± 0.05 per cent of each.

It is good practice to anneal or normalise parts made of this steel at 880±20°C before rough machining, after which they are quenched and tempered. After quenching in oil from 860°± ±10°C and tempering at 600-650°C the specimens cut out of the

test bars (Fig. 126) made of steel Л40Г2 should have the follow-- ing mechanical properties: $\sigma_s \geqslant 40$ kg/mm², $\sigma_b \geqslant 75$ kg/mm², $\delta_s \geqslant 12$ per cent, $\psi \geqslant 30$ per cent, $a_k \geqslant 4$ kgm/mm² and $Bhn = 4.15$-3.85 mm.

After quenching in oil from $860 \pm 10°C$ and tempering at 550-$600°C$ the mechanical properties should be: $\sigma_s \geqslant 60$ kg/mm², $\sigma_b \geqslant 90$ kg/mm², $\delta_s \geqslant 8$ per cent, $\psi \geqslant 15$ per cent, $a_k \geqslant 2$ kgm/cm² and $Bhn = 3.75$-3.50 mm.

The cross section of the ingot for the specimens is X-shaped. This ingot serves to cut out specimens not only for testing the mechanical properties of ordinary steel but also for testing heat-resistant steel for durability and creep at a high temperature. Pieces for fatigue testing of steel are made from ingot 3 (Fig. 126) which is somewhat longer.

The moulding box is made from a piece of iron tube 110-120 mm in diameter and with a wall thickness of 2-4 mm. Holes are drilled in the tube at intervals of 50-70 mm for the passage of gases.

Steel of grade Л40Г2 is typical for the second strength category. It is used in various branches of machine building and for railway rolling stock.

33. COPPER STEEL

Copper increases strength, ductility, impact strength and hardness of steel and corrosion resistance in sea water. In this respect it resembles nickel but costs several times less. Copper steel has a very high hardenability which makes it suitable for heavy castings. At heavy engineering plants use is made of grades 15ДЛ-30ДЛ with 1.25-1.75 per cent C. After normalising at 840°C and ageing at 540°C with a two-hour holding and another tempering at 500°C the mechanical properties of this steel will be as follows: $\sigma_s = 45$-55 kg/mm², $\sigma_b = 60$-70 kg/mm², $\delta_s = 15$-20 per cent, $\psi = 22$-45 per cent, $a_k = 3.5$-6.0 kgm/cm². High-carbon steel of grade 15ОД2СХЛ ($\sigma_s = 50$ kg/mm², $\sigma_b = 70$-75 kg/mm², $\delta_5 = 3.7$ per cent and $Bhn = 280$) is used to cast engine crankshafts.

Unicomponent silicon, chromium or nickel steels are not rational grades for casting since they require large additions of an alloying element to improve noticeably the casting and mechanical properties.

Copper is used instead of nickel in multicomponent steels of various grades to improve the fluidity of moulds and other engineering properties of iron-base and nickel-base alloys (Monel, 30 per cent Cu, and 70 per cent Ni, etc.).

34. SILICO-MANGANESE STEEL

Manganese behaves much in the same manner as nickel—it increases the strength and ductility of steel. Low hardenability does not allow it to be used for thick-walled castings. The steel of grades 15ГСЛ-45ГСЛ is the most popular. After normalising at 840°C or hardening from 840°C and tempering at 550°C, silico-manganese steel has the following mechanical properties: $\sigma_s = 35\text{-}50$ kg/mm^2, $\sigma_b = 60\text{-}70$ kg/mm^2, $\delta_5 = 20\text{-}30$ per cent, $\psi = 45\text{-}55$ per cent, $a_k = 8\text{-}12$ kgm/cm^2. This steel is employed to make parts for agricultural and printing machines, small parts for railway cars, guide rollers of overhead cranes, etc. Silico-manganese steel has a higher strength and ductility than carbon steel. Its low cost makes it extremely attractive to designers and foundrymen who wish to make the structures lighter. Taking the yield point of this steel as equal on an average to 40 kg/mm^2 and that of carbon steel equal to 25 kg/mm^2 it can be seen that, given the same strain, the loaded section of a casting made of steel grade 25ГСЛ will be 1.6 times lighter as will the weight of the entire part.

The steel of grade 20ГСЛ is used to cast blades for hydraulic turbines weighing up to 25 tons and surfaced with sheets of stainless steel as well as many parts for light, medium and heavy machines. The steel readily yields to welding but has a poor fluidity and cracking resistance.

35. MULTICOMPONENT ALLOY STEELS

In multicomponent alloy steels the crystal lattice of iron is distorted more than in one- or two-component steels which increases the strength of the castings. Such steels belong to the highest strength categories. But this is not enough to estimate finally the strength of the castings in the given operating conditions. More data as to the hardenability and temper brittleness, which may cause the failure at ordinary temperatures, are required as well as knowledge of other engineering properties which should be studied most attentively.

The sensitivity of steel to temper brittleness can be conventionally represented as

$$\frac{a_{k_1} - a_{k_2}}{a_{k_1}} = 100\%,$$

where a_{k_1} = impact strength of steel cooled after tempering in water, kgm/cm^2;

a_{k_2} = impact strength of steel cooled after tempering in a furnace at a rate of 10°/hr, kgm/cm^2.

Temper brittleness is typical not only of chrome-nickel, but also of chrome-manganese, nickel-manganese and nonalloyed steel.

The temper brittleness increases as the temperature of Mesnager test pieces drops. At a negative temperature temper brittleness is known as cold shortness. A. Zavyalov believes that the positive effect of molybdenum in preventing temper brittleness consists in the fact that it delays the precipitation of carbides of chromium and manganese from the solution and also of phosphides which cause temper brittleness. Molybdenum carbides which
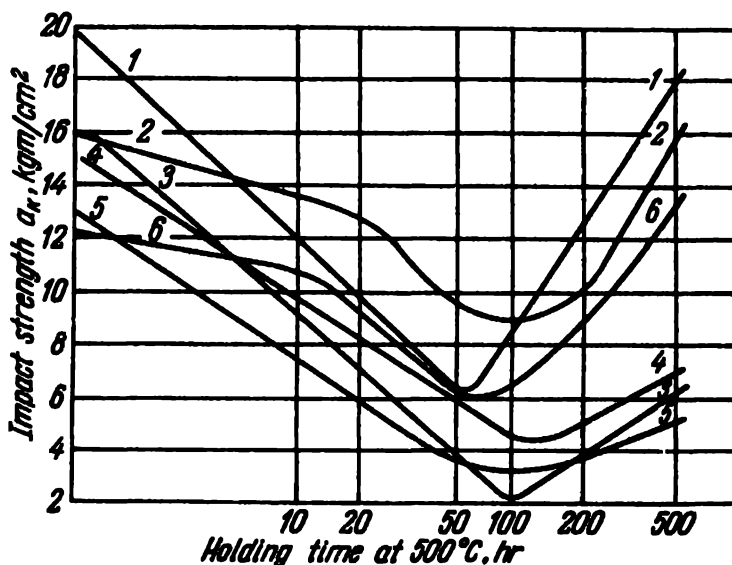
Fig. 127. Kinetics of temper brittleness in various alloy steels (Prosvirin, Kvashnina):

1—grade 40X steel; *2*—grade 35ХГ steel; *3*—grade 35ХГС steel; *4*—grade 35ХГН steel; *5*—grade 35ХГ2 steel; *6*—grade 35ХН steel

failed to dissolve completely at hardening from a normal temperature serve as solidification centres for large carbide masses and prevent the formation of fine carbides—the sources of temper brittleness.

Temper brittleness of cast steel is characterised by the impact strength in steel test pieces subjected to isothermal heating at 500° C after improvement, as compared with the impact strength observed in the test pieces after a prolonged heating from 10 to 500 hr. Embrittlement is most pronounced after heating to 500° C for 100 hr (Fig. 127).

The tendency of steel towards temper brittleness can be also determined at low temperatures by the drop of the impact strength curve. The phenomenon of decrease in impact strength is known as brittleness threshold or the threshold of cold shortness (Fig. 128).

In grade 35ХГВ steel tungsten tends to decrease the impact strength and reduce cold shortness, shifting the threshold of cold shortness to the right (Fig. 128).

Tungsten content, per cent	Temperature of cold shortness threshold, °C
None	40
0.8	60
1.25	80

The temper brittleness and, hence, embrittlement and cold shortness decrease as the content of molybdenum is raised to 0.5 per cent and that of tungsten to 1.6 per cent. When the content of these elements is increased further. the temper brittleness will be more pronounced.

The following conclusions can be drawn from the above.
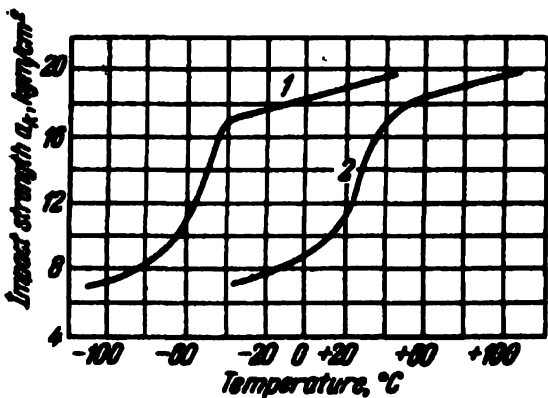
Fig. 128. Threshold of cold shortness of grade 35ХГ steel in the cold (Prosvirin, Kvashnina):

1—tempering at 650° C. two-hour cooling in water; *2*—same and tempering at 500° C, 50 hours

In selecting a structural steel for castings which are to operate at impact load, preference should be given to a steel in which embrittlement is reduced to the minimum (it can never be avoided completely) or which responds weakly to embrittlement. After normalising, important castings should be subjected to high tempering and rapid quenching in water.

Titanium and niobium tend to reduce slightly the embrittlement of steel. On the other hand, molybdenum and, especially, tungsten considerably increase the strength of steel at impact loads both at the ordinary and low temperatures.

The properties of widespread high-strength multicomponent grades of steel conform to an appropriate State standard. The properties of expensive grades of such steel as provided by American foundrymen are illustrated in Tables 33 and 34.

After normalising from 920° C and tempering at 500° C with subsequent quenching in water the steel of grade 18ХНВЛ containing 0.17 per cent C, 0.32 per cent Si, 0.57 per cent Mn, 0.95 per cent Cr, 2.31 per cent Ni and 1.39 per cent W has the follow-

Table 33

Chemical Composition of Steel, per cent

Grade	C	Mn	Si	Ni	Cr	Mo
25 ГХМЛ	0.28	1.36	0.41	1.18	0.68	0.36
35 ГХМЛ	0.34	1.58	0.40	1.22	0.71	0.32
40 ГХМЛ	0.42	1.48	0.34	1.26	0.60	0.39

ing mechanical properties: σ_s=84.3 kg/mm^2, σ_b=98.3 kg/mm^2, δ_5=16.3 per cent, ψ =48.7 per cent, Bhn=269, a_k=16.3 kgm/cm^2.

The steel of this composition is intended for parts which are to operate at impact load.

Investigations of the mechanical properties depending on the wall thickness of cast test pieces have shown that for all the above grades of steel and heat treatment procedures the yield point is reduced on an average by 10 per cent, ultimate strength by 6-7 per cent, elongation by 15-18 per cent and reduction of area by 25-30 per cent, when the wall thickness is increased from 12 to 100 mm. This change in the mechanical properties is directly dependent on the physical homogeneity of the thin and on the heterogeneity of heavy sections. A heavier section favours the formation of liquation films on the surface of large metal grains as well as shrinkage microporosity between them.

The data cited above allow us to make the following conclusion: a carbon steel casting for unique heavy parts is most irrational. The best mechanical properties are observed in castings made of low-alloy steel with a high hardenability.

The heavy engineering industry widely uses grade 30ХГСЛ steel to cast parts subjected to shocks and wear*, grade 30ХМЛ for parts of high-pressure boilers and grade 18ДГСЛ for the blades of hydraulic turbines and other hydraulic units.

Besides good strength and ductility, some grades of multi-component steel possess a high wear resistance, grade 30ХГСЛ, for example. After normalising from 880-900° C and tempering at 400-430° C, this steel containing 0.3 per cent C, 1.2 per cent Mn, 0.7 per cent Si and 0.7 per cent Cr will have the following properties: $\sigma_s \geqslant$ 35 kg/mm^2, $\sigma_b \geqslant$ 63 kg/mm^2, $\delta_s \geqslant$ 14 per cent and $\psi \geqslant$ 30 per cent. This steel is used to make reducing gears of coal crushers (weighing about 900 kg) on which low-pitch herringbone teeth are cut subsequently.

A simpler steel of grade 30ХГСНТ with 0.3 per cent C, 1.2

* The parts are normalised and tempered.

Table 34

Mechanical Properties of High-strength Multicomponent Steel

Grade	Heat treatment	σ_s, kg/mm²	σ_b, kg/mm²	δ, %	ψ, %	Bhn	a_k, kgm cm²
25 ГНХМЛ	900° C, 2 hr, air, 370° C, water	108.8	128.0	11.5	32.8	364	2.2
	900° C, 2 hr, air, 540° C, water	82.0	119.8	13.5	31.2	364	6.3
	900° C, 2 hr, air, 675° C, water	65.0	82.0	22.2	51.6	255	20.2
	900° C, 2 hr, air, 840° C, 4 hr, oil, 540° C, water	125.5	133.0	12.7	33.3	402	9.7
	900° C, 2 hr, air, 640° C, 2 hr, oil, 675° C, water	67.2	82.0	22.0	45.7	255	18.6
35 ГНХМЛ	900° C, 2 hr, air, 370° C, water	127.8	175.0	9.2	29.7	477	2.8
	900° C, 2 hr, air, 540° C, water	103.6	128.4	13.5	36.6	387	8.4
	900° C, 2 hr, air, 675° C, water	67.2	90.7	22.7	51.2	269	18.7
	900° C, 2 hr, air, 840° C, 2 hr, oil, 540° C, water	124.8	143.0	11.0	24.1	418	7.8
	900° C, 2 hr, air, 840° C, 2 hr, oil, 675° C, water	71.7	92.2	19.5	45.7	269	19.0
40 ГНХМЛ	900° C, 2 hr, air, 370° C, water	122.0	150.0	10.0	31.5	430	2.6
	900° C, 2 hr, air, 540° C, water	106.5	129.2	14.5	36.2	402	7.3
	900° C, 2 hr, air, 675° C, water	66.5	82.0	19.5	42.7	269	21.0
	900° C, 2 hr, air, 840° C, 2 hr, oil, 540° C, water	123.4	141.0	10.7	25.3	418	6.3
	900° C, 2 hr, air, 840° C, 2 hr, oil, 675° C, water	68.7	91.4	19.2	36.4	269	19:0

per cent Mn, 1.1 per cent Si, 0.3 per cent Cr, 0.5 per cent Ni and 0.03-0.07 per cent Ti has better mechanical properties ($\sigma_s \geqslant$ 50 kg/mm^2, $\sigma_b \geqslant 73$ kg/mm^2, $\delta_s \geqslant 10$ per cent, $\psi \geqslant 20$ per cent and $a_k \geqslant 5$ kgm/cm^2) because normalising is replaced in this case by quenching from 830°C in air and tempering at 650°C with subsequent cooling in air. It is used to make such small and simple parts as rolls, disks and brackets.

Bit cutters for boring oil wells are cast from steel of grade 25X2ГН2Д2Ф (~ 0.25 per cent C, $\sim$ 1 per cent Mn, $\sim$ 2 per cent C, $\sim$ 2 per cent Cu, $\sim$ 2 per cent Ni and $\sim$ 25 per cent V). The steel is dense and uniform in structure and possesses good casting and mechanical properties ($\sigma_s \approx 140$ kg/mm^2 and $a_k \approx 8$ kgm/cm^2) as well as better impact strength, when case-hardened, in air or in a liquid corrosive medium.
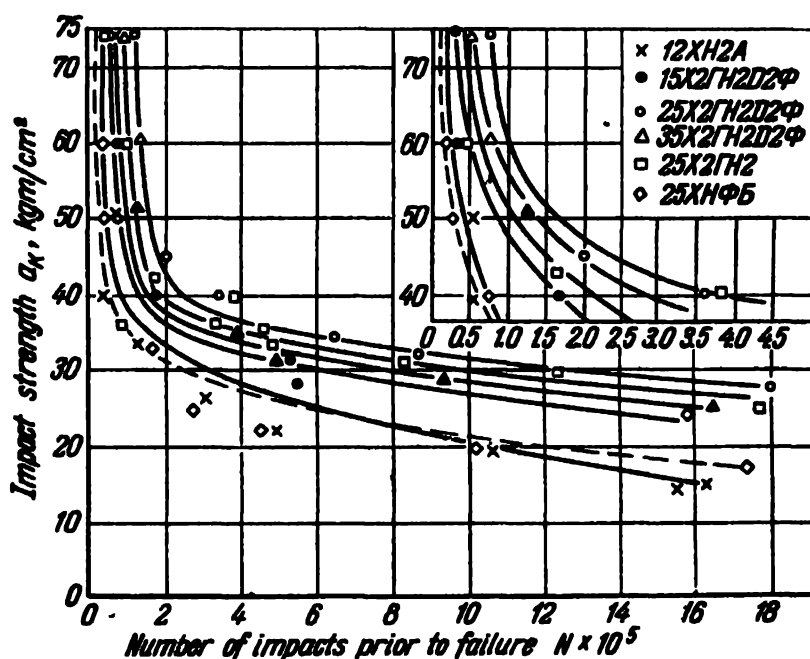
Fig. 129. Impact endurance of multicomponent low-alloy steels (Nekhendzi)

The impact endurance of steels of the grades compared is illustrated in Fig. 129. The maximum impact endurance is displayed by case-hardened steel of grade 25X2ГН2Д2Ф and simple steel of grade 32X2ГН2Д2Ф. A bit made of the former grade of steel can bore granite to a depth 35 per cent more than a bit made from ordinary steel of grade 12XH2A.

To obtain steel of the third strength category without additions of nickel or molybdenum, K. Grechin, N. Tuchkevich and others have developed a steel of grade Л35XГСА (Chromansil). It is intended for important castings. The chemical composition of this steel should be within the following limits: 0.32-0.40 per

cent C, 0.6-0.9 per cent Si, 1.0-1.2 per cent Mn, 0.7-1.0 per cent Cr, $\leqslant 0.40$ per cent Ni, $\leqslant 0.030$ per cent S, $\leqslant 0.035$ per cent P, ~ 0.1-0.2 per cent V.

Long ago, more reliable castings made of structural multi-component grades of steel of the highest strength categories began to oust castings made of aluminium and magnesium alloys. These steels are melted in electric arc and induction furnaces. Important castings are made of two grades—Л35ХГСА and Л27ХГСНА. The chemical composition of the former grade is: 0.32-0.40 per cent C, 0.6-0.9 per cent Si, 1.0-1.2 per cent Mn, 0.7-1.0 per cent Cr, $\geqslant 0.40$ per cent Ni, $\geqslant 0.030$ per cent S, $\geqslant 0.035$ per cent P and 0.1-0.2 per cent U.

Before rough machining the castings should be annealed or normalised at 920 $\pm 20°$C for 2 hours.

With respect to mechanical properties as determined on the pieces cut out from test bars the steel should satisfy the following requirements:

(a) after quenching from $890\pm10°$C in oil and tempering at 630-670°C: $\sigma_s \leqslant 60$ kg/mm^2, $\sigma_b \leqslant 80$ kg/mm^2, $\delta_s \leqslant 12$ per cent, $\psi \leqslant 30$ per cent, $a_k \leqslant 4$ kgm/cm^2, $Bhn = 4.05$-3.75 mm;

(b) after quenching from $890\pm10°$ C in oil and tempering at 570-630°C: $\sigma_s \leqslant 85$ kg/mm^2, $\sigma_b \leqslant 100$ kg/mm^2, $\delta_5 \leqslant 8$ per cent, $\psi \leqslant 20$ per cent, $a_k \leqslant 2.5$ kgm/cm^2, $Bhn = 3.6$-3.35 mm.

The steel of grade Л35ХГСА represents the third strength category.

The steel of grade Л27ХГСНА with an addition of nickel belongs to the fourth or fifth strength category. The chemical composition of this steel lies within the following limits: 0.24-0.30 per cent C, 0.5-0.8 per cent Si, 0.9-1.2 per cent Mn, 0.7-1.0 per cent Cr, 1.4-1.8 per cent Ni, $\leqslant 0.035$ per cent S and $\leqslant 0.035$ per cent P.

The mechanical properties of the steel should satisfy the following requirements:

(a) after quenching from $890\pm10°$C and tempering at 200-240°C: $\sigma_s \leqslant 120$ kg/mm^2, $\sigma_b \leqslant 150$ kg/mm^2, $\delta_s \leqslant 6$ per cent, $\psi \leqslant 20$ per cent, $a_k \leqslant 2.5$ kgm/cm^2, $Bhn = 3.0$-2.70 mm;

(b) after quenching from $890\pm10°$C and tempering at 280-320°C: $\sigma_s \leqslant 100$ kg/mm^2, $\sigma_b \leqslant 130$ kg/mm^2, $\delta_5 \leqslant 7$ per cent, $\psi \leqslant 25$ per cent, $a_k \leqslant 2.5$ kgm/cm^2 and $Bhn = 3.2$-2.85 mm.

The steel is analysed as to its chemical composition according to an appropriate State standard.

The test bars for testing the chemical composition are taken during pouring. No less than two tests are taken from each heat (Fig. 126). The tensile test is performed on a test piece of any diameter with the design length $l : d = 5$. The impact strength should conform to a state standard.

36. HIGH-ALLOY STEELS WITH SPECIAL PROPERTIES

In some cases steel products must meet special requirements which cannot be found in structural alloy steel and more so in carbon steel—namely a high resistance to wear and to a chemically active environment (alkalis or acids, chemically active flue gases containing sulphur compounds, atmospheric moisture), high resistance to destruction due to the action of water vapours, etc.

In other cases the products must possess adequate strength and ductility at high or, conversely, low temperatures. These properties of steel are known as heat resistance and cold resistance.

The requirements of the first group are made to the surface of the product and of the second to the metal.

A separate group includes steel with electromagnetic properties which goes to make permanent magnets, parts of transformers, electric motors and other machines with a high magnetic induction and permeability.

It sometimes happens that steel even abundantly alloyed with one or several elements fails to satisfy all the special requirements. In this case use is made of high-temperature alloys with cobalt, uranium and other elements as their bases, cemented carbides or cast basalt (basalt casting).

According to the classification suggested by the French foundryman L. Guillet in the early 20th century, which is still in use today, high-alloy, mainly stainless, steels are subdivided into three classes depending on their structure.

The first class includes martensitic steels containing 0.1-0.4 per cent C and over 12-16 per cent Cr. They are hardened as the usual grades and possess high strength and especially high hardness.

Ferritic steels with 16-30 per cent Cr belong to the second class. They fail to harden during cooling. They resist oxidation better that class one steels.

Austenitic steels are in the third class. When heated or cooled they retain austenite and do not become stronger on hardening. This class is represented by steel with 18 per cent Cr and 8 per cent Ni, grade Г13Л steel and high-temperature alloys with nickel or cobalt as their bases as well as more complex steels not infrequently with 10 or 12 alloying elements.

The fourth class comprises pearlitic steels which are not included in the Guillet classification and which have a high fatigue strength at ordinary and elevated temperatures. The steel of this class contains several alloying elements. It has come into use not so long ago.

Steel of any class may be characterised by one or several special surface or volumetric properties, which give the steels their names in conformity with their purpose.

Stainless steel effectively resists atmospheric corrosion. Scale-resistant steel has a higher resistance to scale formation and, as a rule, operates at small loads. In addition to the resistance to scale formation, heat-resistant steel must preserve its strength at variable loads and high temperatures. Cold-resistant steel has a high impact strength at low temperatures (minus 40 to minus 100°). It follows from this that, as distinct from stainless steel which must possess only specific surface properties, heat- and cold-resistant steels have both specific surface and volumetric properties.

Steel used for cutters and other tools in the metal cutting machines should be sufficiently strong and hard at red heat. This property is referred to as red hardness.

The chemical resistance of steel in acids and alkalis ranks among the surface properties. Depending on its purpose the steel of this type is called acid-resistant, alkali-resistant, etc. Steel with 4-6 per cent Mo effectively resists the action of a weak hydrochloric acid. Both in the Soviet Union and many other countries this steel is called antichlor. The simultaneous resistance of steel to the chemical action of gases dissolved in water and to the impact from gas bubbles is termed cavitation which takes the form of wear on the blades of hydraulic turbines.

There are steels with magnetic and nonmagnetic properties as well as steel with a high coercive strength, transformer steel, steel for permanent magnets, etc.

37. AUSTENITIC MANGANESE STEEL

When steel contains about 10 per cent manganese it exhibits a purely austenitic structure since manganese widens the region of γ-iron solutions. In steel containing from 10-13 to 20-25 per cent Mn the phase of γ-iron precipitates together with δ-phase having a hexagonal tightly packed lattice.

V. Lipin of Russia was the first to study the behaviour of manganese in steel in 1885. Later, Robert Hadfield continued in England these studies and discovered after three years of research that hardening sharply increased the ductile properties of steel with 12 per cent Mn. The content of elements in Hadfield steel varies within the following limits: 0.9-1.5 per cent C, 0.3-1.2 per cent Si, 10-14 per cent Mn, 0.08-0.15 per cent P and 0.01-0.04 per cent S. Products made from this steel have a high wear resistance especially if their surfaces are cold hardened. The castings made from this steel are machined at low cutting rates using tools tipped with superhard alloys.

Steel with <1.25 per cent C, < 0.08 per cent Si, > 12.5 per cent Mn, <0.12 per cent P and about 0.02 per cent S has the following mechanical properties after hardening: $\sigma_s = 25\text{-}40$ kg/mm².

$\sigma_b = 80\text{-}100$ kg/mm², $\frac{\sigma_s}{\sigma_b}100 = 30\text{-}40$ per cent, $\delta_{10} = 40\text{-}55$ per cent, $\psi = 35\text{-}45$ per cent, $a_k = 20\text{-}30$ kgm/cm² (by Mesnager), $\sigma_{-1} = 15\text{-}20$ kg/mm² and $Bhn = 180\text{-}220$.

Grade Г13Л steel is deoxidised by adding 0.03-0.04 per cent Al and 0.03-0.05 per cent Ti. At the "Serp i Molot" Iron and Steel Plant in Moscow this steel is deoxidised with a mixture consisting of 40 per cent slaked lime and 60 per cent Al powder.

Y. Nekhendzi added 3-4 per cent Al to grade Г13Л steel and obtained a steel with totally nonmagnetic properties (even without quenching in water) which could be effectively machined with an ordinary cutting tool at a small speed and feed.

To make cast steel ductile it is good practice to reduce sharply its carbon content (to 0.25 per cent) and change the content of manganese by the following formula developed by Y. Nekhendzi

$$\% \text{ Mn} = 13.5 + 14\% \text{ C.}$$

Grade Г13Л steel possesses a high wear resistance due to cold hardened surface layer, i. e., more compact crystalline lattice of the austenite. Externally, cold hardening manifests itself in the fact that Brinell hardness of steel after hardening does not exceed 150 (ball diameter—10 mm, load—3,000 kg). If the indentation hardness is measured by a ball of a smaller diameter it will reach 400-450. This explains why the teeth and lips of an earth digger bucket made from this steel can be used to handle gravel for about a year without replacement and wear out completely in several weeks when employed to handle soft soil.

Sudden failure is the most essential shortcoming of castings made of Г13Л steel. Brittleness can be apparently explained by the presence of residual carbides which had no time to dissolve when the steel was heated for hardening. Residual carbides can be frequently observed when the castings are underheated or when hardened at too low a temperature as, for example, when part of the products was cooled in air. Of secondary importance are the insufficient manganese-to-carbon ratio (below 10) and the saturation of the metal poured into a ladle with manganese oxides. The pouring is done at a temperature below 1350°C.

After a suitable holding in a furnace and after all carbides have dissolved in the austenite the castings are quenched in water. The temperature should not be below 1150°C. When the holding time is too short the carbides will not dissolve. The drop in the temperature of the castings held in the air will cause the dissolved carbides to precipitate again from the solid solution. This wrong procedure will sharply reduce the ductility and wear resistance of steel.

Castings from steel Г13Л were produced for the first time at the "Serp i Molot" Iron and Steel Plant in Moscow. In 1927, N. Mel-

nikov and the author succeeded in casting tramway spider frogs and their import from England was stopped in the thirties. This steel is now widely used in mass production of caterpillar tracks. In the U.S.S.R. about 0.5 million tons of tracks are cast every year. In the United States similar tracks are made by rolling. Austenitic steel of grade Г13Л finds an extensive application in mining, especially in grinding equipment, as the cones of ore crushers, and jaws for stone crushers and beaters, plates and balls for coal mills, etc. It should be noted that the balls and plates made from cast basalt are much superior in wear resistance to castings made from manganese steel and cost much less. Balls of high-chromium pig iron cast in metal moulds are less expensive than their steel counterparts but sometimes have a poorer wear resistance.

Grade Г13Л steel is used on a vast scale to make buckets, lips and teeth of earth diggers, dredges and gold mining machinery. The great variety of castings made from this steel is due to its high casting properties: good fluidity and mould filling ability for thin-walled and large-size castings. The casting process is attended by a number of difficulties, however. Thus, linear shrinkage of Г13Л steel is 3 per cent, i. e., one and a half times more than in carbon steel. All measures should be taken, therefore, to prevent casting cracks, improve the pliability of the moulds, etc. This feature is specific of the production techniques used to obtain grade Г13Л steel. The resistance to cracking is promoted by the low pouring temperatures, good deoxidation of steel with complex deoxidisers and other measures. Castings made without risers may develop porosity.

Grade Г13Л steel is melted at 1350°C and poured into the moulds at 1400°C. Superheat in pouring into moulds should be avoided as much as possible in the case of heavy castings.

Each plant follows its own procedure in filling the moulds. The methods used in casting caterpillar tracks cannot be recommended, for example, in the production of excavator buckets. The plants producing similar products of steel Г13Л must pool their experience to overcome the shortcomings which may cause sudden failure in operation.

Grade Г13Л steel contains a fairly large amount of carbon which is dissolved in a liquid metal. The solubility of carbon decreases on cooling. The excess of carbon in the form of the complex carbide $m\mathrm{Fe_3C} \cdot n\mathrm{Mn_3C}$ gradually precipitates from the solution and settles on the boundaries of the austenite grains during solidification. As-cast steel becomes brittle.

When the castings are heated and held at 1150°C the carbides gradually dissolve in solid austenite. To dissolve the carbides completely the castings should be held at the required temperature for one hour per each 25 mm of their thickness. To prevent

the carbide precipitation from the solution on another cooling the castings are immersed into cold water to quench them. A purely austenitic structure without carbides can be obtained, firstly, after holding the castings sufficiently long at the proper temperature and, secondly, after hardening them from the temperature not below 1125°C. If these requirements are not observed the carbides will fail to dissolve completely. As a result, the austenite grains will be separated by a brittle substance which will disintegrate even under light blows and the castings will break.

The machinability of austenitic steel can be improved by increasing the manganese content to 17-20 per cent and decreasing the content of carbon to 0.20-0.25 per cent. This steel can be easily machined and is nonmagnetic. It is used to cast washers for magnetic cranes operating in bedding plants and headworks. Annealed castings are slowly cooled in the furnace. The manganese-to-carbon ratio should be within 7.5-10.

Today, extensive research is underway to improve the casting methods used for steel Г13Л and improve the operating properties of wear-resistant steels in general.

38. STAINLESS STEEL

Stainless steel is highly resistant to corrosion in the atmosphere and other aggressive media (acids, alkalis, hot gases which contain disulphide, sea water, etc.).

Stainless steel goes to make blades for steam and hydraulic turbines, parts of pumps, shafts of ships, propellers, bearings, journals, springs, kitchen utensils, surgical instruments, parts of measuring tools, etc.

The most popular is the steel alloyed with chromium or with chromium and nickel, manganese, silicon, molybdenum, titanium, etc.

In all concentrations chromium produces a solid solution of γ-iron but narrows down its region (Fig. 130). Fifteen per cent of chromium is the maximum amount at which steel retains its austenitic structure.

When steel contains 44-60 per cent Cr it develops intermetallic phases which belong to FeCr. Chromium dissolved in cementite produces duplex carbide $(Fe, Cr)_3C$. A higher percentage of chromium will yield carbide $(Fe, Cr)_7C$ and carbide $(Fe, Cr)_4C$ in a low-carbon steel. Carbide $(Fe, Cr)_{23}C_6$ is present in steel containing very little carbon.

Various structures may be obtained in chromium steel depending on the content of carbon (see Fig. 131), because chromium changes the mechanical and physical properties of steel. Being a carbide-forming element, chromium increases the hardness and strength, yield point and impact strength without detracting anything

from the ductile properties in the presence of carbon. Normalising sharply increases the strength of the castings.

On the surface of steel castings and ready products chromium quickly oxidises and forms a solid and strong film of oxides which protect the metal from further disintegration.

The corrosion resistance of chromium steel depends on the content of carbon and the heat treatment procedure. The more carbon there is in the steel and the less chromium, the sooner it will rust (Fig. 131). Thus, in selecting the composition of stainless steel it will be well advised to increase the content of chromium and decrease the concentration of carbon.
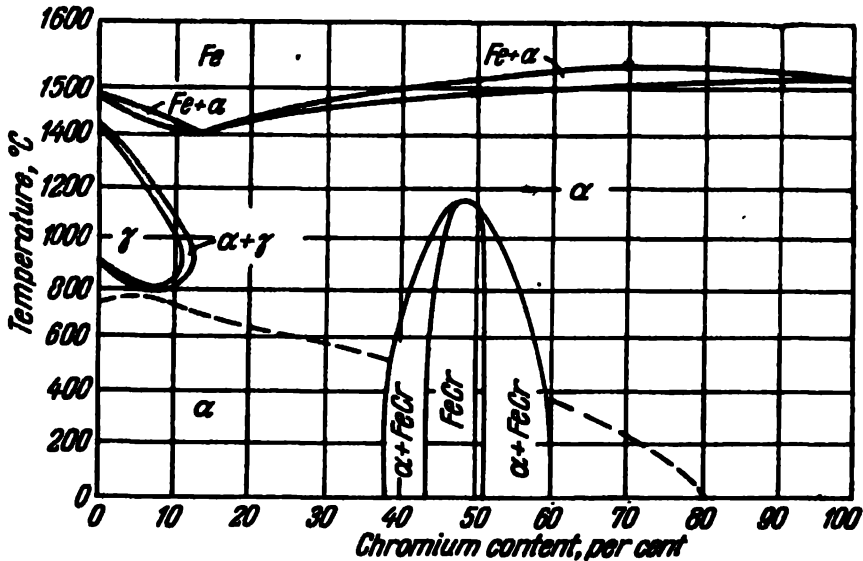
Fig. 130. Constitutional diagram of binary alloys Fe-Cr

A layer of scale on the surface of the product formed by chromium increases the scale resistance of the castings. The limiting temperature of scale resistance depends on the structure of the metal and its carbon content, in which case the mechanical properties of steel castings are of secondary importance.

When high-carbon steel is slowly heated within the range of 600-800°C a brittle intermetallic σ-phase precipitates. This process is accompanied by a considerable reduction in the volume of the product which gives rise to internal stresses and cold cracking and decreases sharply the ductility of the metal making it brittle. For this reason, castings should be cooled as fast as possible in this temperature range (600-800°C), and then slowly in the furnace to relieve the internal stresses. When the castings made of steel with 15-18 per cent C are cooled slowly within the 400-550°C range they again develop temper brittleness, which American foundrymen call the 475°C brittleness. Since this complicates the casting procedure a steel of a more complex composition is ordinarily used.

Chromium promotes the growth of the primary steel grains and the size of the primary crystals cannot be changed by heat treatment in the alloys without the phase transformation $\gamma \rightarrow \alpha$. In the case of a slow solidification the grains become larger in proportion to the wall thickness of the casting. The heterogeneous structure of the castings made of high-chromium steel causes dissimilar mechanical properties and corrosive strength of the specimens cut out from the various places in the same casting with different wall thicknesses.

The primary grains are refined by introducing ferrochromium, saturated with nitrogen, and ferrotitanium into the steel.
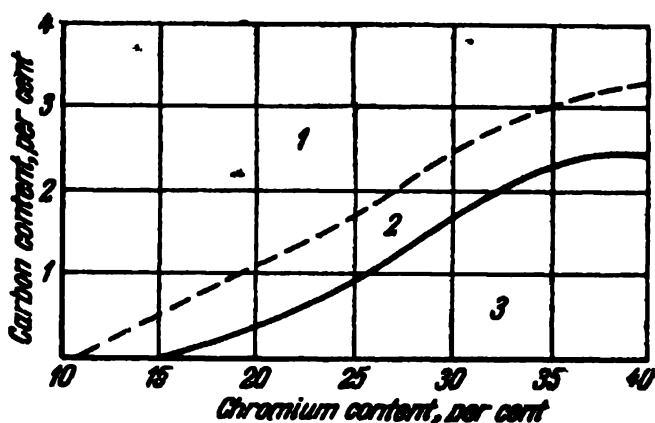
Fig. 131. Change in the steel resistance to corrosion depending on carbon and chromium content (Roesch):

1—rust; 2—no rust after heat treatment; 3—no rust without heat treatment

Chromium increases the acid resistance of steel castings in strong inorganic acids — sulphuric and nitric, but not in hydrochloric acid. The acid resistance can be improved by adding 2-3 per cent Ni to the steel, besides 16-20 per cent Cr. This steel belongs to the martensitic class ($\sigma_s = 60$ kg/mm^2, $\psi = 30$-40 per cent). Phase transformations take a complete course in this steel used mainly in the chemical industry.

In its properties, ferrite-carbidic steel is similar to hypereutectoid carbon steel while ledeburitic steel resembles hypoeutectic pig irons.

Purely ferritic chromium steel exhibits a higher resistance to corrosion than carbidic or ledeburitic steel. Carbides deteriorate the machinability of steel and for this reason the composition of the steel should be selected from the diagram in Fig. 131. It is usual practice to use chromium steels in the region 1.

The latest classification of anticorrosive steels and other alloys places them among the following five main groups.

1. Martensitic steels with 12-17 per cent Cr and 0.1-0.5, sometimes up to 1 per cent C. On rapid cooling from high temperatures they are hardened.

2. Ferritic steels with 16-30 per cent Cr. In a steel with less chromium the carbon content should be at its minimum (0.35 per cent in steel with 30 per cent Cr). This steel fails to harden but it exhibits a tendency towards grain growth at a high temperature and towards brittleness.

3. Austenitic steels with chromium, nickel and other elements. At ordinary temperatures their austenitic structure is preserved. This group is represented by the widely-known steel with 18 per cent Cr and 8 per cent Ni, although the nickel content may be increased to 35-40 per cent. These steels tend to develop intercrystalline corrosion.

4. Austenite-ferritic steels containing a small amount of ferrite in addition to austenite. They yield to deformation and are sensitive to intercrystalline corrosion. The steel ordinarily used in industry contains 0.5 per cent C, 0.5 per cent Si, 0.5 per cent Mn, 20-22 per cent Cr, 8 per cent Ni, 2.5 per cent Mo and 1.5 per cent Cu, or 0.05 per cent C, 0.8 per cent Si, 3 per cent Mn, 20 per cent Cr, 2 per cent Ni and 1.5 per cent Cu.

5. Alloys containing no or little iron, nickel and cobalt base. Almost all of them have an austenitic structure. The group includes alloys resistant to aggressive media at a low or moderate temperature. The chemical composition of martensitic stainless steels subdivides them into four groups (Table 35).

Classification of Martensitic Stainless Steel

(Colombier and Hochmann)

Group	Content of, %		
	C	Cr	Ni
I	Less than 0.15	12-14	—
II	0.20-0.40	13-15	—
III	0.6-1.0	14-16	—
IV	0.1	16-18	2

The ability to harden is advantageous in two respects: it increases the resistance to corrosion and the steel becomes so strong as to be on a par with ordinary structural steels. As the carbon content is increased in the steel, part of the chromium is expend-

ed to form carbides. To preserve the anticorrosive properties of this steel the used part of the chromium is compensated for by an appropriate increase in its content.

The steel of the first group is quenched from 1000°C in water (small parts) or in oil (heavy parts). The higher is the tempering temperature (500-750°C), the greater will be the elongation and impact strength and the less the strength, yield point and even hardness.

Tempering at 250°C is quite enough to relieve internal stresses.

The steel of the second group goes to make decorative articles and table sets. Its strength is improved by quenching in oil from 1000°C.

The steels of the first and second groups are sensitive to temper brittleness. An addition of 2-3 per cent Si reduces the ability of these steels to harden due to the formation of ferrite. Manganese stabilises the austenite, while copper improves the resistance to corrosion.

The steel of the third group is used to make especially hard articles. Since the heat conductivity of this steel is not high, the castings should be heated slowly to avoid thermal stresses and cracking. The steel becomes especially strong when hardened from 1050°C. Shaped parts should be normalised at the same temperature.

The steel of the fourth group is used in place of marine bronze. This steel resists effectively electrochemical corrosion. The hardening temperature should not exceed 1000°C. The mechanical properties of this steel as well as of the steels of the other groups are almost identical: $\sigma_b \approx 120$ kg/mm^2, $\sigma_s \approx 80$-100 kg/mm^2, $\delta_s = 11$-15 per cent, $a_k = 7$-12 kgm * (impact strength). The steel of the fourth group exhibits a better corrosion resistance than the steels of the other groups, especially in organic acids (formic, citric, tartaric, lactic, malic and oxalic).

Steel of the ferritic class undergoes no transformations on heating or cooling.

The steel of the first group should contain not less than 0.12 per cent C and 15-18 per cent Cr. It has a higher resistance to corrosion than martensitic steel, especially in sea water and in nitric and organic acids. The specific gravity of this steel is 7.6 and that of martensitic steel 7.7-7.75 . The steel of the second group contains 25-30 per cent Cr. Prolonged holding within the range of 500-700°C causes the formation of the intermetallic phase FeCr or the so-called σ-phase.

Poor weldability of ferritic steels containing chromium limits the sphere of their application.

* Mesnager test piece.

Small additions of nickel (up to 2 per cent) or copper (up to 1 per cent) tend to improve the corrosion resistance and fluidity. Hardening somewhat increases the strength of the steel: σ_b=55-60 kg/mm^2, σ_s=37-42 kg/mm^2.

An addition of 0.5 per cent Ti markedly improves weldability but slightly reduces the strength without affecting ductility. Molybdenum improves the resistance to corrosion in acetic acid.

Low-carbon ferritic steel is very difficult to cast because the pouring temperature has to be increased, which promotes the grain growth. For this reason, the carbon content in steel is increased
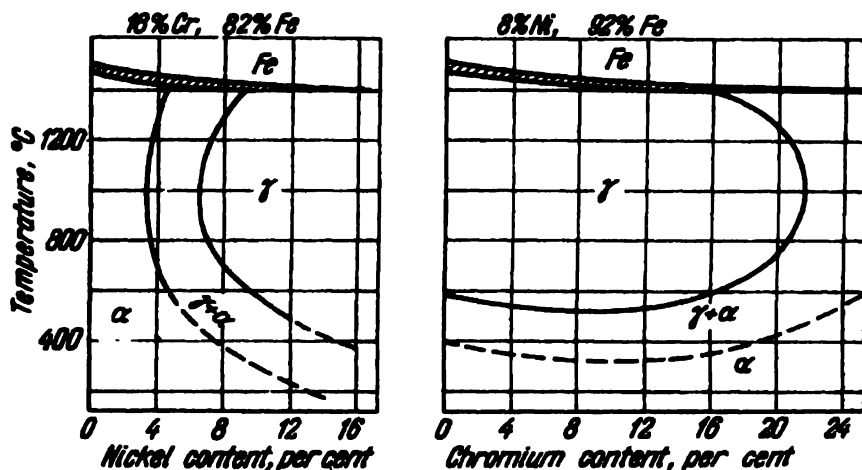
Fig. 132. Effect of nickel on the structure of steel with 18 per cent Cr (left) and the effect of chromium on the structure of steel with 18 per cent Ni (right) (Nekhendzi)

to 0.2-0.3 per cent and that of silicon to 0.7-1.0 per cent. The castings are tempered at 850°C or normalised.

Steel with 25-30 per cent Cr has a purely ferritic structure even after hardening from very high temperatures except for the case when very thin sections are quenched in water. Since the steel has a tendency towards temper brittleness, the castings annealed at 800-900°C are normalised and cooled in the air or water. The steel is highly sensitive to notching. Notched samples have an extremely low impact strength. As the temperature is increased the steel grows more ductile. For this reason, the castings made from this steel can be used at higher temperatures not exceeding, however, 800°C. Heating above 950°C causes an intercrystalline corrosion.

This steel goes to make parts which are to operate mainly in gaseous media containing sulphurous compounds.

Austenitic steels rank among the alloys which contain two or more components since chromium alone produces a very limited austenitic region.

Steel with 18 per cent Cr and 8 per cent Ni is most popular. It contains less than 0.12 per cent carbon, and sometimes even less than 0.03 per cent. When added in small amounts, titanium and niobium prevent intercrystalline corrosion.

The constitutional diagram of Fe-Ni-Cr alloys (Fig. 132) shows that about 7 per cent Ni is required to obtain an austenitic structure after quenching in water from 900-1150°C.

When the chromium content exceeds 22 per cent the steel is no longer purely austenitic but acquires an austenite-ferritic
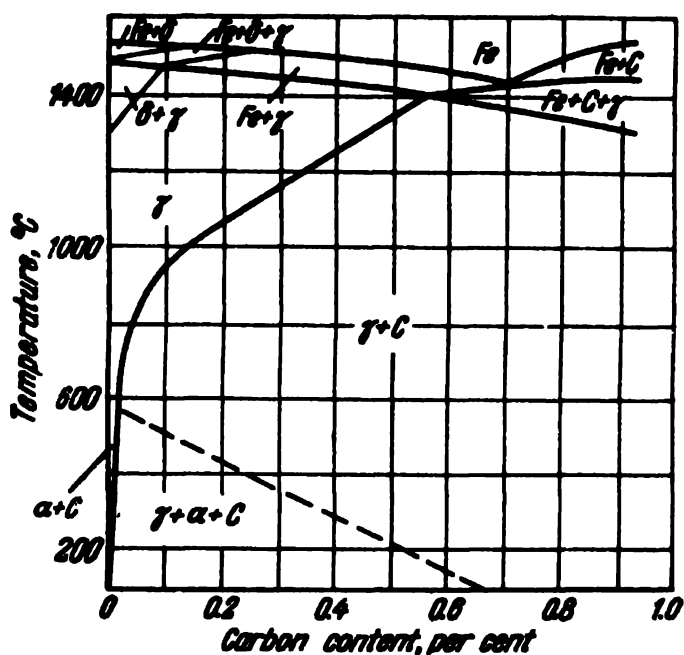
Fig. 133. Pseudobinary constitutional diagram of the alloys Fe-Ni-Cr-C (for section of 18 per cent Cr and 8 per cent Ni) and the effect of carbon on steel structure (Nekhendzi)

structure. In type 18-8 steel which contains 0.001 per cent N after remelting in vacuum a ferritic structure can be obtained even by quenching in water from 1150°C.

Under the usual conditions of melting in an electric furnace the steel contains 0.02-0.04 per cent N and has an austenitic structure. This is very important in selecting the composition and structure of the steel in conformity with the purpose which the castings are intended to serve.

The shortcoming of chrome-nickel and nickel-chromium high-alloy grades of steel is their tendency towards intercrystalline corrosion which causes the products to break on impact like cardboard. Devoid of intergranular bonds, the steel grains easily separate from each other. The steel loses its strength and ductility. Intercrystalline corrosion is caused by the precipitation of carbides at a definite composition and temperature (Fig. 133). The

critical points are shifted to the left here: the eutectic point has shifted from 0.3 to 0.7 per cent C and the concentration point E from 2 to 0.5 per cent C. Steel with over 0.7 per cent C is a regular pig iron. Steel with 0.5-0.7 per cent C belongs to the austenite-carbidic group and corresponds to the ordinary steel with 0.8-2.0 per cent C.

Carbides begin to precipitate from the solution at 600°C when the carbon content ranges between 0.02 and 0.5 per cent. To stabilise the carbides in the solution the castings should be quenched in water, the quenching temperature increasing with higher carbon content. If the carbon content is 0.2 per cent the quenching is done at 1050°C. With 0.3 per cent carbon in the steel the temperature is raised to about 1200°C.
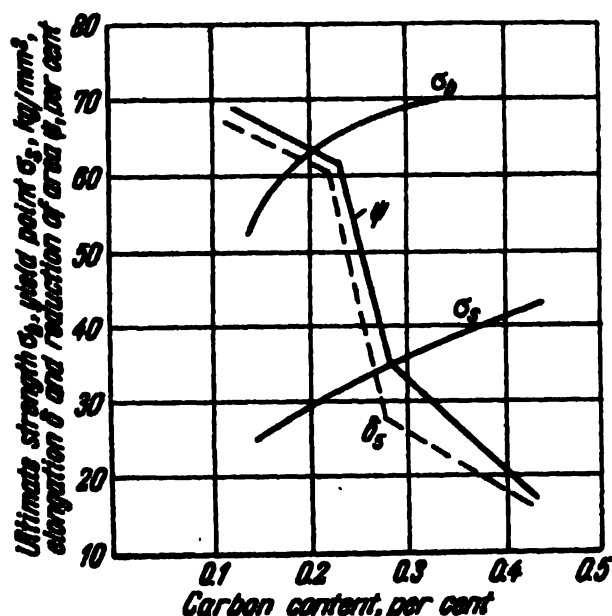
Fig. 134. Change in the mechanical properties of cast steel of grade X18H8 after quenching from 1050° C in water as a function of carbon content. The castings are test strips 45 mm thick, cast in a sand mould (Nekhendzi)

During hardening, the castings made of steel X18H8 develop internal stresses since this steel has a low heat conductivity. But cold cracking is quite rare because this steel is very ductile. The internal stresses must not be relieved by heating because the carbides will segregate on the grain boundaries even after holding for 3 minutes at 700°C. Heavier castings made from steel X18H8 must contain less carbon which must not exceed 0.12-0.16 per cent.

Fig. 134 illustrates the effect of carbon on the mechanical properties of steel with 18 per cent Cr, 8.8-9.1 per cent Ni, 0.81-1.22 per cent Si and 0.54-0.74 per cent Mn (strips 45 mm thick, cast in sand, after quenching from 1050°C). The impact strength of steel is very high (25-30 kgm/cm³), the resistance to bending fatigue is 25 kg/mm², i. e., about 0.55 σ_b, provided the carbon content is not over 0.07-0.12 per cent. Castings of medium thickness

must be made from steel with 0.08-0.16 per cent C. The hardening temperature can be decreased to 1000-1050°C.

The almost inevitable precipitation of the carbides is prevented by introducing into the steel stronger carbide-forming elements: titanium (0.4-0.5 per cent) or niobium (0.8-0.9 per cent) which bind the carbon in the form of carbides which precipitate from the solution, not so fast as chromium carbides.

Titanium deoxidises the steel and refines the primary grains but sharply reduces its fluidity. Less expensive ferrochromium with 0.3-0.4 per cent C may be added in the presence of titanium.

For important castings made from scale-resistant steel the content of chromium is increased up to 25 per cent and that of nickel to 20 per cent, the amount of carbon not exceeding 0.18 per cent.

The parts are cast in dry-sand moulds, in centrifugal machines or by the investment casting method. After the risers and gates are removed the castings are heated to 1050-1100°C, held at this temperature for 1-3 hours and cooled in water or air. After annealing and normalising the mechanical properties should be as follows: $\sigma_b \leqslant 50$ kg/mm^2, $\delta_s \leqslant 25$ per cent, $\psi \leqslant 30$ per cent, $Bhn = 142$-187 kg/mm^2.

Low-carbon steel of grade X25H20T is melted in an ordinary induction furnace. When steel is melted in electric arc or open-hearth furnaces the carbon content in steel at the end of melting can be reduced with a great difficulty and the process requires a long period of decarburisation. However, in the presence of oxygen the final melting period can be considerably shortened.

When added in amounts of 2-3 per cent, silicon tends to improve the scale resistance and fluidity. The manganese content should be decreased to 0.4-0.5 per cent to avoid surplus carbon and phosphorus.

Successful experiments have been conducted for the first time to replace the expensive nickel by manganese. Steel with 18 per cent Cr, 8 per cent Mn and 0.1 per cent C has found application in industry. Only low-carbon ferromanganese is used. This is a two-phase steel (50 per cent of austenite and 50 per cent of ferrite).

To obtain a purely austenitic structure, the content of nitrogen should be increased to 0.2 per cent and that of chromium to 20-22 per cent, in the presence of 3.5 per cent Mn and 3.5 per cent Ni.

The stabilisation of the austenite in scale-resistant steels is improved by adding tungsten, molybdenum, titanium, cobalt, niobium and other elements which increase both scale and heat resistance. One such steel is grade 21-11-2.5. Its chemical composition varies as follows: 0.10-0.25 per cent C, 0.7-1.5 per cent Si, 0.6-1.2 per cent Mn, 20-22 per cent Cr, 11.0-12.5 per cent Ni, 2.4-3.0 per cent W, 0.05-0.20 per cent Ti, $\leqslant 0.25$ per

cent Mo, $\leqslant$ 0.2 per cent V, $\leqslant$ 0.04 per cent S and $\leqslant$ 0.045 per cent P. The castings are normalised at 1050-1100°C for 3-8 hours. Castings which are to operate in strenuous conditions and at a high temperature are also subjected to ageing (heating to 800-850°C, holding for 5-10 hours and cooling in the air).

The mechanical properties should meet the requirements indicated in Table 36.

<div align="right">*Table 36*</div>

Mechanical Properties of Cast Steel of Grade 21-11-2.5

Test tempera-ture, °C	Heat treatment procedure	Mechanical properties, not below			
		σ_b, kg/mm²	δ_5, %	ψ, %	*Bhn*, mm
20	Normalising at 1050-1100° C	50	15	20	4.3-5.0
20	Same plus ageing at 800-850° C	50	12	14	4.3-5.0
700	Same	20	18	—	—

At high temperatures the mechanical properties are determined when required.

The silchrome alloy is used as a scale-resistant steel. Its chemical composition varies within very broad limits. Silchrome contains 2 to 3 per cent silicon. The scale resistance of the castings can be increased by adding more aluminium. There are also castings made from fechral (0.25 per cent C, 13-15 per cent Cr and 3.0-4.5 per cent Al) and chromal (0.20 per cent C, 30 per cent Cr and 5 per cent Al) which exhibit good scale resistance within the temperature range of 700-1250°C, an open-grain structure and high brittleness. The steel of these types has poor fluidity, tends strongly towards oxide-film formation and has a low resistance to cracking. Fluidity can be improved by increasing the content of nitrogen and raising the pouring temperature which weakens the oxide-film formation. On strong superheating, however, the solidification of the metal is attended by a strong shrinkage and the resistance to hot cracking is sharply reduced. For this reason, foundry practice assigns for the steel of each grade definite temperature ranges and pouring rates at which the amount of casting defects can be reduced to the minimum. The more carbon and chromium there is in the steel, the lower should be the superheat.

High-chromium steel contracts within a wide range of solidification temperatures and the shrinkage cavities become more

dispersed. Therefore, if a high hermeticity is not required of the castings it suffices to provide for thickened sections on the gates, as in the case of malleable pig iron castings, dispensing with the risers altogether. The metal should be poured at high rates.

Internal chills do not fuse in chromium steel and are therefore not used.

In the process of producing castings from high-chromium grades of steels with a low resistance to cracking at the solidification temperatures, a number of complex measures should be taken to eliminate hot cracking. Castings made of high-alloy steel are very sensitive to cold cracking, which also requires special measures to be taken in the production process.

39. IRON-BASE AND OTHER HEAT-RESISTANT ALLOYS

Castings made from stainless or anticorrosive steel should possess adequate corrosion resistance in aggressive media at elevated temperatures and under load and a minimum of special mechanical properties. Heat resistance is a term applied to describe the resistance to small plastic deformations at elevated temperatures.

Obviously, each stainless or anticorrosive steel can be heat-resistant and, conversely, a heat-resistant steel can be both stainless and anticorrosive.

Heat-resistant alloys are differentiated by the temperatures at which they are intended to operate. In the casting industry these alloys are conventionally classified into three groups: one covering those installed in services involving temperatures below 400°C, the other between 400 and 700°C and the third above 700°C.

The structure of heat-resistant steel can be pearlitic or austenitic while the high-temperature steel may contain a small amount of iron or made with nickel, cobalt or other than iron bases. The structure of a heat-resistant alloy must be sufficiently stable to preclude the loss of strength and embrittlement. Prolonged operation of the alloy at a high temperature may cause the formation and growth of crystals of new phases which come out from the solid solution. The solid solution is impoverished by corresponding alloying admixtures and loses its original properties while the new phase grows, coagulates and disturbs the bonds between the primary grains. In physical metallurgy this process of transformation is known as ageing.

Heat-resistant alloys are evaluated on the basis of creep at a fixed temperature or stress at which the specimen elongates by 1 per cent during the specified time expressed in hours, for example 5.5 kg/mm² at 500 and 100,000 hours. Creep can be observed, for instance, in bolts which fasten together the flanges

of a steam conduit in a steam turbine when their length gradually increases and the loosened nuts have to be periodically re-tightened. A steel specimen subjected to a creep test gradually elongates and becomes narrower until it breaks in its thinnest section. The problem of creep has been the subject of extensive studies in the U.S.S.R. and other countries.

Fine-grain metal in which the volume of grains is relatively small will display a higher resistance to creep during deformation and, conversely, if the creep is determined by the deformation of the intergranular substance a higher resistance to creep will be typical of coarse-grain alloys. Practical experience shows that a coarse-grain metal has a better stability at high temperatures.

The effect of addition of one per cent of various elements on the creep of low-carbon steel with 0.2 per cent C at 500°C for 50,000 hours is illustrated in Table 37 after hardening from 1100-1300°C and tempering at 400°C.

Table 37

The Effect of Addition of One Per Cent of Various Elements on the Creep of Low-carbon Steel at 500° C

(Bennek and Bandel)

Alloying element	Increase in the creep limit in as-hardened steel compared to as-annealed steel, per cent	
	Bainite structure	Martensitic structure
Niobium	300	230
Titanium	200	130
Vanadium	175	110
Molybdenum	130	70
Tungsten	100	20
Aluminium	40	70
Manganese	30	Not below 10

Above 500°C the carbides and other phases which determine the resistance to creep coagulate and the bainite and martensitic structures change into a ferrite-pearlitic structure. This completely nullifies the advantages accruing from alloying. At 550°C and above and under relatively small loads the annealed structure exhibits a better resistance to creep than the hardened

structure—bainite or martensite. At 600°C the austenitic structure of multicomponent steel proves the best.

In France, England and the U.S.A., the steel of the following composition is used for steam turbines with high parameters (550-600°C) (L. Colombier and J. Hochmann):

C	Si	Mn	Cr	Mo	V	W
0.20	0.3	0.5	0.2	0.65	0.26	—
0.17	0.3	0.5	3.0	0.50	0.70	0.50
0.10	0.3	0.5	1.5	1.0	0.50	—

For steam turbines with capacities of 150 and 200 thousand kW, Soviet foundrymen use steel of grade ЛА-1 (austenitic cast steel) of the following composition: $\leqslant 0.16$ per cent C, $\leqslant 0.55$ per cent Si, $\leqslant 0.70$ per cent Mn, $\leqslant 0.02$ per cent S, $\leqslant 0.025$ per cent P, 14-16 per cent Cr, 2.8-3.2 per cent Co, 1.8-2.2 per cent Mo, 1.8-1.2 per cent W and 0.15-0.35 per cent Ti. This steel is distinguished by good fluidity. The low pouring temperature causes blowholes, oxide films and spider-like twists and the high temperature—shrinkage porosity, casting cracks, etc.

High-strength alloys go to make parts which are to operate at a temperature above 700°C (vanes for jet engines). By their chemical composition these alloys are subdivided into three groups:

(a) complex iron-base austenitic alloys;
(b) nickel-base alloys in which iron is an admixture;
(c) cobalt-base alloys.

The strength of these alloys is due to the formation of intermetallic phases which strengthen the solid solution because of the distortion of the crystal lattice during the period preceding the emergence of these phases. Niobium and titanium are introduced into the alloy, besides molybdenum, vanadium or tungsten, to promote the effect of the formation of the intermetallic phases. The carbon content of 0.2-0.4 per cent in high grade alloys brings about undesirable consequences since the carbides that form tend to reduce the resistance to creep. In these alloys cobalt retards coagulation of the intermetallic phase and the carbides.

Table 38 illustrates the chemical composition of highly heat-resistant steels containing more than 20 per cent Fe and used in some countries.

The creep limit at an elongation of 1 per cent during 1,000 hours for the Sirius HT alloy is shown in Table 39.

Nickel-base alloys were developed for the first time by the Monde Nickel Co. on the basis of nichrome 80-20. Additions of 1.8-2.7 per cent Ti and 0.5-1.0 per cent Al into the alloy noticeably increased its resistance to creep. Several grades of the alloy called Nimonic appeared which were used for the vanes

Table 38

Highly Heat-resistant Steels Used in France, England and U.S.A.
(Colombier and Hochmann)

Country	Alloy	Chemical composition, %						
		C	Cr	Ni	Co	Ti	Nb	Others
France	ATU7	0.05	18	30	25	2	—	—
	Sirius HT	0.20	17	14	10	2.0	—	3W
	ASR	0.30	18	11	11	—	—	10W
England	918	0.40	13	13	10	—	3	2Mo, 2W
	932	0.27	10	10.5	47	—	1.4	2.2Mo, 3W
U.S.A.	Timken	0.10	16	25	—	—	—	6Mo, 0.15W
	No. 155	0.15	20	20	20	—	1.0	3Mo, 2W, 0.15N
	S588	0.45	20	20	—	—	4	4Mo, 4W
	S590	0.45	20	20	20	—	4	4Mo, 4W
	Refractaloy 70	0.05	20	20	30	—	—	8Mo, 4W
	Discaloy 24	0.03	13.5	26	—	1.8	—	3Mo

Table 39

Creep of Sirius HT Steel
(Colombier and Hochmann)

Heat treatment	Creep kg/mm² at test temperature, °C		
	700	750	800
Hardening from 1200°C	14	9	5.5
Same and 15-per cent cold hardening	18	13	7.5

of aircraft jet engines as well as cobalt-base alloys, of the "vitallium" type used in the U.S.A. (Table 40).

Nimonic 90 (0.08 per cent C, 0.5 per cent Si, 0.40 per cent Mn, 58 per cent Ni, 20 per cent Cr, 16 per cent Co, 2.5 per cent Ti, 1.6 per cent Al and 0.5 per cent Fe) and Nimonic 95 (0.08 per cent C, 0.5 per cent Si, 0.40 per cent Mn, 58 per cent Ni, 20 per

Table 40

Chemical Composition of Nickel-base
Alloys

Alloy	Content per cent											
	C	Si	Mn	Cr	W	Mo	Co	Ti	Al	B	Fe	Ni
Nickel	0.1	1.5	1.0	—	—	—	—	—	—	—	—	The rest
Nimonic:												
75	0.1	0.5	0.5	20	—	—	—	0.4	0.2	—	0.8	Do
80	0.05	0.5	0.7	20	—	—	—	2.3	1.0	—	0.5	Do
90	0.10	0.5	0.5	20	—	—	16	2.4	1.2	—	0.5	Do
D	0.10	0.5	1.0	18	—	—	—	—	—	—	The rest	37
DS	0.10	2.0	1.0	18	—	—	—	—	—	—	Do	37
Niconel	0.10	0.5	0.5	16	—	—	—	—	—	—	7	The rest
Corronel												
B	0.05	0.5	0.5	—	—	28	—	—	—	—	6	The rest
Waspaloy	0.38	0.75	1.0	19.5	—	4.25	—	2.5	1.25	—	2	Do
GMR235	0.15	0.60	0.25	15.5	—	5.25	—	2.0	0.30	0.06	10	Do
Hastelloy C	0.05	0.50	0.50	15	4.5	17	—	—	—	—	20	Do
Brightray C	0.05	1.4	—	20	—	—	—	—	0.3	0.1	—	Do
Monel	0.10	1.0	1.0	—	—	—	—	—	—	—	30	Do
Vitallium, Haines No. 21	0.27	0.60	0.30	27	—	5.5	64	—	—	—	1.0	2.0
X40; L=251 Haines No. 36	0.40	0.60	0.60	25	14.5	—	5.5	—	—	—	1.0	10.0

cent Cr, 16 per cent Co, 2.5 per cent Ti, 1.6 per cent Al and 0.5 per cent Fe) are characterised by the following rupture strength:

Temperature, °C	Rupture strength, kg/mm^2
815	29
900	10
925	8

After heating at 1150°C (4 hours), cooling in air and heating at 700°C (16 hours) and cooling in air to obtain a high resistance to creep and prevent the grain growth over the broad temperature

range the castings made from these alloys are tested for short-time strength and creep up to 10,000 hours at high temperatures.

On short-time tests Nimonic 95 showed the strength and ductility indicated in Table 41. Table 42 shows resistance to creep at a high temperature.

Table 41

Mechanical Properties of Nimonic 95 at Short-time Temperature Tests

Test tempera-ture, °C	σ_b, kg/mm²	σ_4, %	ψ, %	Modulus of elasticity $E = 10^{-4}$, kg/mm²
20	114.7	25.1	24.1	21.3
500	100.0	23.6	28.1	19.1
700	72.1	4.5	7.1	17.6
900	30.7	10.9	12.5	12.5
1000	8.7	59.3	48.0	9 9

Table 42

Resistance to Creep of Nimonic 95

Tempera-ture, °C	Creep limit, kg/mm² at residual elongation						Ultimate rupture strength, kg/mm²		
	100 hr	5,000 hr	100 hr	10,000 hr	100 hr	500 hr	100 hr	1,000 hr	5,000 hr
750	28	12.9	30.8	11.2	32.2	14.7	32.9	22.4	15.14
815	15.4	—	16.8	—	18.2	—	18.9	11.9	—
870	9.8	—	11.2	—	11.9	—	12.6	7.7	—
925	—	—	—	—	—	—	7.0	3.1	—

Cobalt-base alloys of the stellite type are used in investment pattern casting. The pouring temperature should be very high so as to obtain the highest resistance to creep.

Cobalt-base vitallium alloy composed as indicated in Table 40 was tested for rupture strength for 100 hours and showed the following values of ultimate strength:

Temperature, °C	Ultimate strength kg/mm²
815	15.4
900	11.7
982	6.0

X40 alloy used in the United States has a somewhat higher strength:

Temperature, °C	Ultimate strength, kg/mm²
815	19.9
900	14.7
982	7.9

Casting of heat-resistant alloys based on iron and other elements is a complex process of which we know very little. The
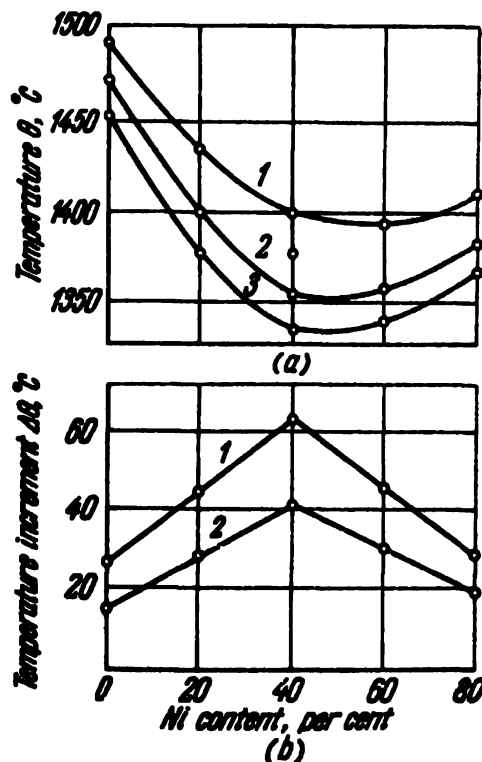

Fig. 135. Change in the temperature range of solidification of nickel alloys depending on nickel content (Nekhendzi):

a—solidus temperature; b—range of solidification temperatures; 1, 2 and 3—steels of various grade

principal properties of heat-resistant castings—rupture strength at working temperature and resistance to creep—depend not only on the chemical composition of the alloy but also on the rates of initial solidification in the foundry mould.

It is of interest that the liquidus temperature of heat-resistant steel and nickel-base alloys remains constant whereas the solidus temperature varies (Fig. 135a) as does the temperature range of solidification (Fig. 135b). This was demonstrated recently by the researchers at the Leningrad Polytechnical Institute. The following conclusion can be drawn: the constitutional diagrams of alloys plotted for the equilibrium conditions of sloweddown cooling cannot be used. New diagrams should be plotted for the respective rates of solidification of metal in the mould. This follows from the fact that the solidification range predetermines the phenomena of shrinkage, the formation of macro- and microstructure of castings and a number of other engineering properties.

40. TOOLS CAST FROM HIGH-SPEED STEEL

The investment casting method brought into use a far broader range of alloy steels. This method gives castings of the required shape and almost completely eliminates subsequent machining by metal cutting tools. Precision casting is the only method for imparting any shape to products made of alloy steels with poor machinability. The advantage of the method is that the cast tools are almost ready for service.

High-speed steel must be able to cut metal at red heat. Classical high-speed steel has the following chemical composition: 0.6-0.7 per cent C, up to 0.05 per cent Si, up to 0.12 per cent Mn, 4.3-5.0 per cent Cr, 17.5-18.5 per cent W and 0.35-0.55 per cent V.

Red hardness is the ability of steel to retain its original strength at 550-650°C. Durable wear resistance of the cutting edges is the main property of high-speed steel. Substitutes for high-tungsten high-speed steels have less tungsten and chromium and more carbon and vanadium.

The following composition is the best for a low-alloy high-speed steel used for cast tools and investigated by A. Dobrovidov and N. Yerofeyev: 0.9-1.1 per cent C, 4-4.5 per cent Cr, 3.5-4.0 per cent W and 2.5-3.0 per cent V. A slight increase in the tungsten content of 4 per cent will improve the cutting properties. As compared to the classical steel 18-4-1, this steel contains far more carbon which forms solid and stable carbides of chromium, tungsten and vanadium with a complex composition; besides, when the tool is heated to the normal working (high) temperature the austenite after hardening changes to martensite which increases the hardness necessary for cutting.

Vanadium deoxidises the steel and forms stable carbides in it. If there is more than 4 per cent vanadium in steel it will impoverish the solution with carbon and decrease the red hardness of the alloy.

The amount of structurally free carbides in steel reaches 12-14 per cent. After triple tempering at 560°C the solid solution contains about 0.7 per cent W, 2.3 per cent Cr and 0.3 per cent V.

According to the procedure developed by N. Yerofeyev, the tools are cast in metal moulds.

I. Mozgovoi suggested his own casting process for bimetallic tools. The shanks are made of carbon steel and the cutting edge of high-speed steel. The tools are cast in a revolving mould. The high-speed steel poured first is thrown by the centrifugal force towards the periphery of the mould where the cutting teeth of the tool are to be cast. Then carbon steel fills the rest of the space in the centre. This method is used to cast slotting cutters, reamers, mills, gear shapers as well as knurling cutters and dies.

Permanent magnets are cast from the so-called permanent magnet alloys in which the residual coercive force magnetism remains high after the magnetisation ceases (Fig. 136).

Although somewhat limited, the problem of obtaining soft magnet castings by metallurgical methods is of tremendous industrial and economic importance.
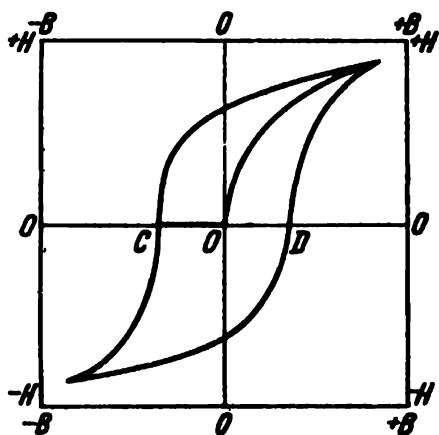
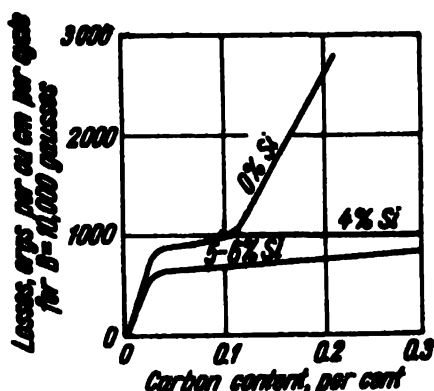*Fig. 136.* Hysteresis loop in magnetising iron alloys

Fig. 137. Change in hysteresis losses depending on carbon and silicon content

Of late, successful experiments were conducted in melting soft magnet steel in vacuum. This made it possible to decrease the carbon content from 0.05-0.06 in ordinary heats to 0.01-0.03 per cent in melting in vacuum.

Armco-iron has the following magnetic properties: $H_c = 0.72$ oersted, $W_h = 2,100$ erg/cm^3 per cycle, $\mu_{max} = 7,000$.

The hysteresis losses are decreased at a higher silicon content (Fig. 137).

With more than 4 per cent silicon in the steel, up to 0.08 per cent C is in the form Fe_3C. When the silicon content exceeds 6 per cent, 0.008 per cent of carbon changes to graphite and silicon no longer affects W_h.

In a transformer steel with 4 per cent Si, W_h can be determined from the equation

$$W_h = 12,000 \, S - 4,000 \, P + 800 \, Mn. \tag{193}$$

Silicon affects the magnetic properties of steel in two ways: it deoxidises and degases the liquid metal and promotes the grain growth, i. e., it behaves as a useful admixture. At the same time, like the other elements accompanying iron, it reduces the limit of saturation magnetisation and magnetic permeability and in this respect is harmful. Silicon is especially beneficial

in that it increases the specific electric resistance thereby decreasing eddy current losses. Depending on silicon content, the specific resistance of steel is expressed by the Gumlich formula

$$R = 0.0999 + 0.12 \, Si \, \varrho \mu \Omega / cm^3 \qquad (194)$$

and, depending on carbon content, by the Yensen equation

$$R = 9.6 + 82.5 \, C + 4.5 \, (C - 0.02). \qquad (195)$$

The Brinell hardness of steel with 0.05-0.08 per cent C and 4.0 per cent Si comes to 220 and that of pure iron to 80. Silicon forms a solid solution in α-iron and when its content is increased to over 4 per cent, the solid solution FeSi can be found in the ferrite. The castings are annealed at 950°C with subsequent stabilisation from 800°C in air.

Steel rapidly gets rusty. Rust formation is decreased in the presence of 0.5 per cent Cu. Frames of electric motors, pole shoes and other such parts in electrical machines are cast from dynamo steel with 0.08-0.10 per cent Cu and 1.5-2 per cent Si.

Magnetic hard alloys must be characterised by a large area of magnetising loop and a big coercive force. In the mid-twenties of this century use was made predominantly of an expensive steel with 30 per cent Co, in which $B = 8,500-9,500$ gausses and $H_c = 220-250$ oersteds. In the early thirties a highly coercive steel containing 10-40 per cent Ni and 5-20 per cent Al made its appearance. This steel had better magnetic properties: $B = 6,000-6,300$ gausses, $H_c = 500-550$ oersteds, but was so hard that it could not be machined even with tools made of superhard alloys.

Thin specimens exhibit higher magnetic properties, which depends apparently on the size of the primary grains and the cooling rate during hardening.

Steel with 28 per cent Ni and 11 per cent Al has the best magnetic properties. After shot blasting and removal of the skin such steel castings can be machined with tools tipped with P9 hard alloy. The castings are annealed at 900°C. To obtain the best magnetic properties the castings are heated to 1230°C, quenched in water (20°C) and tempered at 660-670°C. After such heat treatment the magnetic induction amounts to 6,000 gausses.

Magnets with a wall thickness of 130 mm are cast from steel with 24.3 per cent Ni, 12.7 per cent Al and 3.5 per cent Cu. The cylindrical surfaces and the end-faces of the casting are ground all over leaving the internal surface rough. Then it is hardened from 1100°C in boiling water and tempered at 550°C for 3 hours. This steel was much more resistant to hardening cracks than steel of grade H28IO11.

CHAPTER V

SMELTING STEEL FOR SHAPED CASTINGS

42. SIDE-BLOWN CONVERTER PROCESS

This process consists in refining of molten pig iron used in the production of cast steel, carried out in a side-blown converter where the blast jet strikes the surface of the metal bath. This method is widely used in steel foundries to produce light thin-walled shaped castings because it enables a much more hot metal to be obtained than in blowing through the bottom. Besides, side blowing makes for a better mixing of the metal with slag and oxidised droplets of metal and the molten metal bath absorbs less nitrogen from the air. This has a beneficial effect on the quality of poured metal, guarantees a good fluidity of thin-walled and large-size moulds and increases the ductility of cast steel.

As compared to the open-hearth and electric furnace processes the side-blown converter process has the following advantages:

(1) periodic tapping of hot metal in small amounts as needed by small foundry shops;

(2) low electric power consumption;

(3) high efficiency of the process;

(4) low initial outlay on equipment;

(5) comparatively simple steel smelting operation and rapid training of personnel;

(6) the possibility of obtaining alloy steels by adding ferro-alloys into the ladle.

On the other hand, the side-blown converter process has the following shortcomings:

(1) the need for charge materials comparatively low in sulphur and phosphorus for the acid process;

(2) higher melting losses.

These shortcomings can be eliminated by using oxygen which decreases the content of ferrous oxide in the slag or by replacing the acid lining by basic lining* or else by vacuum treatment of steel in a converter after blowing.

The productivity of small steel making units is illustrated in Table 43.

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* Basic lining is used effectively not only in the U.S.S.R., but also in Austria and the Federal Republic of Germany.

Table 43

Daily Productivity of Steel Making Units
(Kryanin)

Unit	Capacity, tons	Duration of heat, hr	Number of shifts	Number of heats per day	Daily productivity, tons
Service I					
Converter	1.5	0.5	1	16	24.0
Electric furnace . . .	3.0	2.0	2	8	24.0
Open-hearth furnace .	5.0	4.5	3	5.3	26.5
Service II					
Converter	1.5	0.5	2	32	48.0
Electric furnace . . .	5.0	3.0	3	8	40.0
Open-hearth furnace .	10.0	5.0	3	4.8	48.0

Note: Service I — two-shift work.
Service II — three-shift work

The capital outlay for steel making units with the same daily productivity is characterised by the following figures. If we assume the cost of constructing the converter body with a charge capacity of 1.5 tons to equal 100 per cent, then the cost of other units will be (in per cent):

Converter (1.5 tons)	100
Electric furnace (3 tons)	145
Same (5 tons)	175
Open-hearth furnace fired by fuel oil (5 tons) . . .	250
Same (10 tons)	420

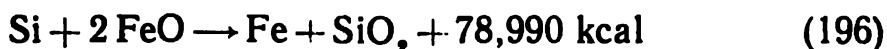According to the actual data, the cost of 1 ton of molten steel obtained from various units will be (in per cent):

Side-blown converter steel	100
Open-hearth steel	112
Electrical steel	146

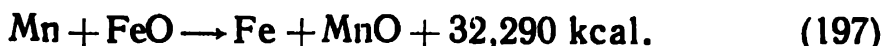The comparative content of nitrogen in steel melted under various conditions (Kryanin) will be, in per cent:

Thomas	0.0155-0.0170
Bessemer	0.0145-0.0185
Electrical	0.0113-0.0160
Open-hearth:	
acid	0.0045-0.0050
basic	0.0040-0.0080
Side-blown converter steel	0.0018-0.0040

The content of nitrogen in steel predetermines its density and cold resistance.

The side-blown converter process consists of the following two clearly defined periods. During the first period the temperature of the bath is rather low, the metal is not covered by slag and the iron is the principal component to be oxidised. The temperature of the bath gradually increases and the silicon and manganese begin to burn at the expense of the oxygen in the ferrous oxide dissolved in the metal
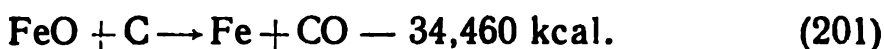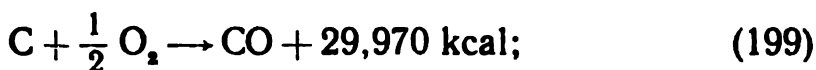
$$Si + 2\,FeO \longrightarrow Fe + SiO_2 + 78{,}990 \text{ kcal} \qquad (196)$$

and

$$Mn + FeO \longrightarrow Fe + MnO + 32{,}290 \text{ kcal.} \qquad (197)$$

Reactions of direct oxidation of iron, silicon and manganese may also occur on the bath surface.

In the temperature range below 1450°C the burning of the carbon is not very well developed as the dissociation elasticity of the carbon monoxide is less than that of SiO_2 and MnO.

When the temperature of the metal exceeds this limit there sets in the second period of melting which is characterised by the burning of carbon and decarburisation of the metal:

$$C + O_2 \longrightarrow CO_2 + 97{,}650 \text{ kcal;} \qquad (198)$$

$$C + \frac{1}{2}\,O_2 \longrightarrow CO + 29{,}970 \text{ kcal;} \qquad (199)$$

$$CO + \frac{1}{2}\,O_2 \longrightarrow CO_2 + 67{,}630 \text{ kcal;} \qquad (200)$$

$$FeO + C \longrightarrow Fe + CO - 34{,}460 \text{ kcal.} \qquad (201)$$

The results of the side-blown converter process in the converter body with an acid lining depend on several factors: the initial silicon content in the cupola pig iron poured into the converter body, its temperature and the initial temperature of heating the converter body lining.

The entire course of the process depends on the heating of the converter brickwork (below 1200°C). The low temperature of heating usually requires much extra silicon because ferrosilicon has to be added during melting to heat the cold bath by the chemical heat of the exothermic reaction (196). Since the first period is delayed excessively much iron burns out, resulting in abundant ferrous slag, and the bath is very difficult to heat because the heat of combustion goes to raise the temperature of the very cold and highly calorific brickwork. In this case, brown smoke begins to appear 15-20 minutes after the air has been supplied. At the end of melting, the steel is cold, which may cause grave accidents in pouring and wholesale rejects due to misruns, oxide

films and blowholes. The strength of the metal is markedly impaired. The normal heating of the converter body is brought to 1350-1400°C.

The principal data of a series of test heats by the three versions of the side-blown converter process depending on the silicon content are represented in Table 44.

A large amount of added silicon tends to increase the melting period by 40 per cent and involves an overconsumption of ferrosilicon up to 35 kg per ton of the metal and is therefore economically unprofitable. Conversely, the low-silicon process (at the usual pig iron superheat) is more rational since the duration of the process is reduced and the output of the converter is increased by 40 per cent. The consumption of the ferrosilicon is reduced to 10 kg per ton of the metal. In this way, a physically hot pig iron with a low silicon content is most advantageous.

Back in 1877, D. Chernov reported to the Russian Technical Society on the results of converting physically hot but chemically cold low-silicon pig iron (with 0.8 per cent Si).

K. Polenov developed a similar process for converting low-silicon pig iron from a reverberatory furnace. This process earned the name of Russian converter process and is distinguished by using physically hot but chemically cold pig iron.

During the converter process the heat is produced by the exothermic reactions of the burning of iron and its admixtures. The reactions of combustion of silicon and carbon in pig iron into carbon dioxide are the most essential sources of heat. This heat is expended to raise the temperature of the cold converter body lining, for radiation and to heat the liquid metal.

The side-blown converter process is distinguished from the bottom-blown process by the burning down of the carbon monoxide into carbon dioxide, which increases somewhat the temperature of the gases above the metal which heats up in this case much better.

Despite these advantages every effort should be made to reduce the consumption of the chemical heat on raising the temperature of the cold lining of the converter body and, especially, after a prolonged idle period of the unit, i. e., the lining should be heated with coke.

When the lining is hot, physically and chemically cold pig iron may be used.

The British converter process is conducted, as is known, on a chemically hot, i. e., high-silicon pig iron, which requires even an artificial cooling of the metal. This is achieved by throwing cold scrap into the converter. At first the process occurs very slowly with the burning out of iron and silicon. The temperature of the gases over the bath is too low and the melting takes a longer time.

Table 44

Comparative Results of the Side-blown
Converter Processes by Three Versions

(Kryanin)

	Process		
	high-silicon	medium-silicon	low-silicon
Chemical composition of cupola pig iron, per cent			
carbon	3.25-3.60	2.98-3.20	2.89-3.31
silicon	1.17-1.78	0.52-0.80	0.60-0.82
manganese	0.55-0.85	0.45-0.79	0.69-0.89
sulphur	0.036-0.056	0.050-0.061	0.051-0.060
phosphorus	0.060-0.030	0.045-0.070	0.060-0.061
Mean temperature, °C:			
of pig iron in pouring	1,260	1,275	1,260
of converter body lining	1,325	1,375	1,395
Mean consumption of ferrosilicon per ton of steel, kg	35	23	10
Total amount of silicon added, per cent . . .	2.05-2.74	1.20-1.58	1.0
Mean duration of the process, min	20	·14	14
Chemical composition of molten steel, per cent			
carbon	0.04-0.10	0.04-0.08	0.02-0.04
silicon	0.010-0.20	0.03-0.28	0.05-0.15
manganese	0.04-0.06	0.04-0.09	0.07-0.14
sulphur	0.034-0.044	0.027-0.047	0.030-0.044
phosphorus	0.052-0.070	0.057-0.0069	0.060-0.070
Temperature of the metal, °C:			
in converter body .	1,720-1,760	1,715-1,760	1,720-1,730
in tapping from the body	1,640-1,720	1,650-1,700	1,590-1,640
in pouring	1,550-1,585	1,560-1,570	1,540-1,570
Chemical composition of the steel after deoxidation · and alloying, per cent			
carbon	0.14-0.20	0.10-0.18	0.16-0.18
silicon	0.32-0.40	0.34-0.40	0.31-0.40
manganese	0.57-1.08	0.59-0.98	0.68-0.84

	Process		
	high-silicon	medium-silicon	low-silicon
sulphur	0.040-0.030	0.044-0.052	0.050-0.062
phosphorus	0.056-0.080	0.045-0.070	0.058-0 068
Mechanical properties of the steel obtained:			
σ_b, kg/mm^2	43.7-49.0	40.6-46.7	44.2-45.8
σ_s, kg/mm^2	24.5-29.6	24.2-29.8	21.6-29.6
δ, %	24.2-30.5	25.6-27.8	27.8-29.0
ψ, %	27.8-51.2	36.0-43.8	39.2-43.9
a_k, kgm/cm^2 . . .	6.6-10.0	6.2-8.2	6.2-8.7
Bhn	127-148	121-143	117-143

I. Kryanin investigated the side-blown converter process on low-silicon and physically hot pig iron and showed its positive aspects. The chemical composition of pig iron varied within the following limits: 2.70-3.2 per cent C; 0.37-0.54 per cent Si; 0.09-0.86 per cent Mn; 0.028-0.045 per cent S and 0.050-0.070 per cent P. The temperature of the pig iron poured into the converter body was 1360-1440°C; and the temperature of the body lining before blowing was 1440-1520°C.

The rise in the temperature of the metal in the converter, the burning of admixtures, the composition of slag and gases in the converter mouth are indicated on the diagrams (Fig. 138). The silicon and manganese burned out from the moment the blast was turned on to the beginning of the seventh minute. The carbon began to burn noticeably from the beginning of the fifth minute. The blowing operation continued for only 14 minutes. The gas contained predominantly CO_2, and not CO. At the seventh minute the gases were found to contain an excess of oxygen (up to 2.5 per cent) which testifies to the hot course of melting. After blowing the high-silicon pig iron, the ductile properties of the metal obtained were, as compared with the ductile properties of steel, higher in the Russian converter process—by 18 per cent in relative reduction of area and by 11.8 per cent in impact strength, which indicates to a higher purity and homogeneity of the metal.

The advantages accruing from the side-blown converter process are determined by the technical and economic indices illustrated in Table 45.

The melting loss of metal during the high-silicon process amounts to 15 per cent and during the low-silicon process to about 12 per cent. In the Russian process there is no need for

ferrosilicon and the duration of heat is reduced almost twofold. The cost of the charge is reduced by 36 per cent per ton of metal. No haematite pig iron is needed. The temperature of pig iron tapped from the cupola should be low, not below 1440-1470°C, i. e., the cupola should be provided with a device for heating the blast.
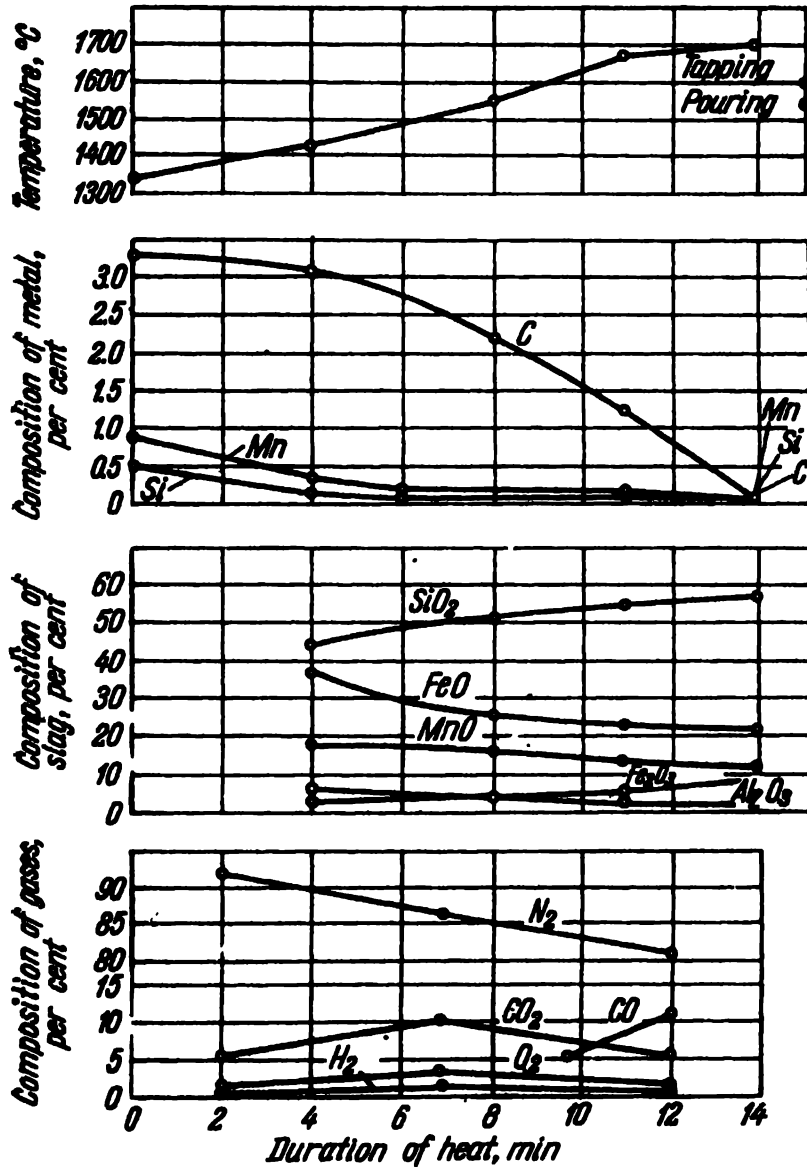

Fig. 138. Diagrams of the Russian side-blown converter process (Kryanin)

A cold converter should be well heated at the beginning of the shift, after repairs or a prolonged idle period. After the steel has been tapped the remaining heat of the converter lining should be preserved by covering the mouth with a lid.

As a rule, the foundries use converters with the charge capacities of 1.5 and 2.5 tons (Fig. 139). The detachable shell makes

Table 45

Comparison of the Data of the Material and Fuel Balance in Two Side-blown Converter Processes

(Kryanin)

	Process	
	British (high-silicon)	Russian
Chemical composition of liquid pig iron, per cent		
carbon	3.60	3.27
silicon	2.05	0.52
manganese	0.85	0.86
Temperature, °C:		
of pig iron poured into the converter	1,260	1,400
of steel being tapped	1,700	1,700
Theoretical yield of good steel, per cent	90.65	94.58
Amount of slag, per cent	10.44	4.72
Heat input, per cent:		
with pig iron	29.8	43.3
with air	8.2	7.4
from carbon oxidation	35.2	39.7
from silicon, manganese and iron oxidation	26.2	9.3
from slag formation	0.6	0.3
Heat consumption, per cent:		
lost with steel	45.9	55.3
lost with slag	7.1	3.7
lost with gases	28.2	26.5
lost to decompose H_2O . . .	1.7	1.4
lost by radiation and convection	17.1	13.1
Useful heat, per cent	53.0	59.0

it possible to prepare the lining of the converter body and to dry it on a special stand and then quickly replace the shell with the used lining by another spare shell with reclaimed refractory material.

The converter is lined with Dinas brick of the 1st class, with the refractoriness 1710°C and the normal size (65×123×250 mm). Fig. 139 shows a converter with a combined lining. The mouth.
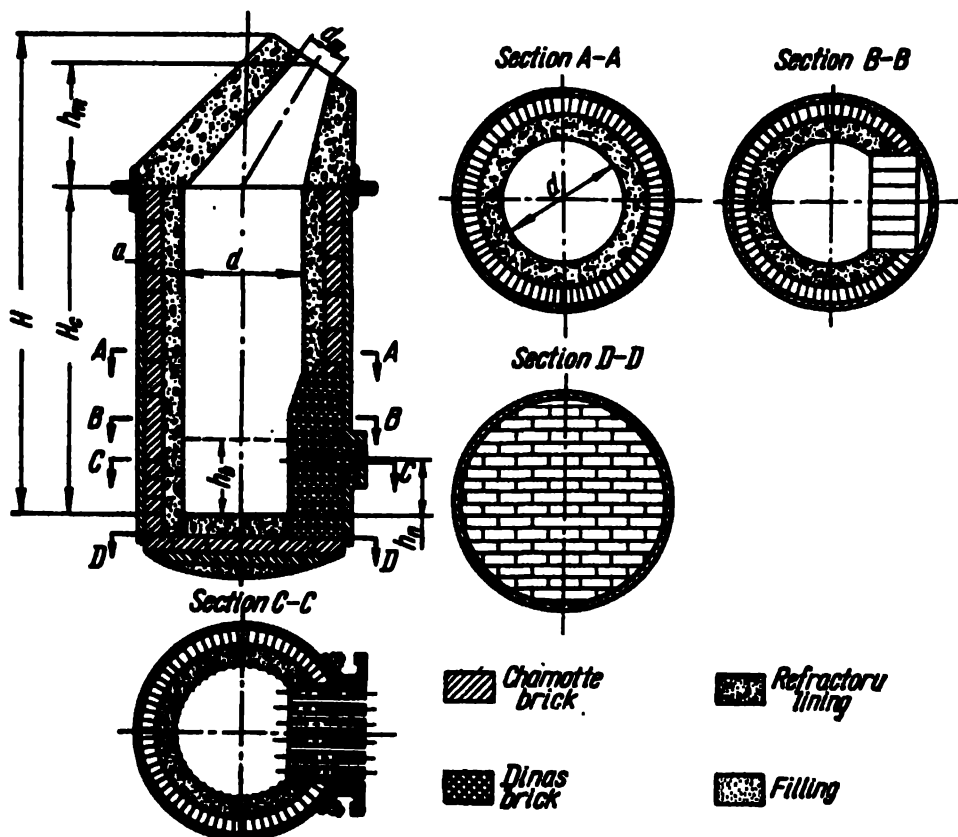

Section A-A   Section B-B

Section D-D

Section C-C

▨ Chamotte brick.   ▤ Refractory lining

▨ Dinas brick   ▨ Filling

Fig. 139. Side-blown converter

and the internal part of the wall are rammed in with a compound to the depth equal to one course of Dinas brick. The composition of the compound is illustrated in Table 46.

The mean consumption of refractories per ton of steel amounts to about 40-50 kg. When the brick are hewed the seams are too thick (about 8-10 mm) and the brickwork disintegrates rapidly. Rammed-in lining serves for a longer time. The thickness of the lining in the area at the tuyeres should be larger than the thickness of the converter body lining by 420-400 mm. The brickwork at the bottom should be not less than 350-400 mm.

For the hot melting process, according to B. Andreyev, the tuyeres are arranged at an angle of 15° to the horizontal. The openings of the tuyeres are usually lined with a wooden form.

The average life of the converter body lining is 100 blows. Corrosion is usually observed at the tuyeres and higher. The

Table 46

Composition of the Compound for Lining the Converter

(Ladyzhensky and Tunkov)

Compound	Composition of dry mass, per cent (by volume)	Fineness of grinding, mm	Amount of water per cu m of dry mixture, lit
Chamotte, thick	Chamotte powder 60-70 Refractory clay 30-40	up to 3	400
Chamotte, half-thick	Chamotte powder 60-70 Refractory clay 30-40	up to 2	500
Dinas, half-thick	Dinas powder 85-90 Refractory clay 10-15	up to 2	500
Lining compound	Ground quartz 60 Refractory clay 27 Chamotte powder 3 Fine coke 8 River sand 2	up to 7 up to 4 up to 3 up to 4	6%
Same	Quartz sand 80-82 Refractory clay 20-18	—	7-8%
Same	Ground quartzite 80-82 Refractory clay 20-18	—	7-8%

endurance of the lining decreases at a higher content of manganese in the pig iron.

The main dimensions of standard converters with the charge of 1.5 and 2.5 tons are shown in Table 47 and in Fig. 139.

The total useful section of all the nozzles per ton of weight of the charge should be within 30-40 sq cm. The consumption of air for blowing 1 ton of pig iron should not be below 800 cu m. The operating capacity of the blower is calculated from the formula

$$Q = \frac{WT}{n} \qquad (202)$$

where Q=consumption of air per ton of pig iron, m³/min;
W=blower capacity, cu m;
T=weight of the metal in the converter, tons;
n=duration of blowing, min.

At the blast pressure of 0.2-0.4 atm the velocity of air fed to the converter with the charge of 1.5 tons should amount to not less than 75 m³/min and to the converter with the charge of 2.5 tons to not less than 100 m³/min.

Table 47

Characteristics of Standard Converters

	Charge weight, tons	
	1.5	2.5
Inner diameter of the body, d, mm	850	1,000
Mouth diameter, d_m, mm	280	300
Height of the body, H, mm	2,900	3,200
Height of the cylindrical part of the body, H_c, mm	2,120	2,370
Height of the mouth, h_m, mm	980	830
Height of the cupola hearth to the axis of the nozzles, h_n, mm	350	450
Thickness of the lining of the cylindrical part, a, mm	250	250
Number of nozzles	6	6
Nozzle diameter, mm	32	40
Maximum cross-sectional area of the body, sq m	0.567	0.785
Area of the metal bath surface, sq m	4.5	7.15
Bath depth, h_b, mm	410	490
Cross-sectional area of all nozzles, sq cm . . .	48.3	75.4
Volume of the converter, cu m	1.40	2.13

After ramming in the lining the converter body is "seasoned" for 6-12 hours and then dried in the flame of gas burners or fire wood for 12-15 hours.

For heating, the converter is filled with coke to 2/3 of its height, the coke is ignited and the body is dried at natural draught with the tuyeres open. The air is then forced periodically into the furnace under a pressure not over 0.2 atm, adding coke. A repaired converter is heated for 8-10 hours to the white heat without yellow spots. An operating converter body is heated for 3-4 hours.

To pour in pig iron the vessel is tilted, the molten metal is poured in through the mouth, the blast is turned on and the converter is returned to the working position.

One or two minutes after the blast is put on, brown smoke comes up from the mouth and then a dark flame which becomes gradually lighter in colour. The length of the flame depends on the burning intensity of the carbon. At first the flame is short and then increases slowly to 3-4 metres.

The colour and length of the flame are the main characteristics of the temperature of the blowing process. Brown smoke

belches from the mouth during the burning of iron and silicon. At the cold starting of the heat the flame may come up only for 6-8 min and sometimes later. Then a shaft of sparks breaks out from the mouth which shows that carbon begins to burn in a low-silicon pig iron. Soon afterwards the sparks change into a dark flame which indicates to the beginning of burning of carbon and afterburning of CO into CO_2 and also to the end of the first and the start of the second melting period.

Towards the end of the second period, i. e., the end of carbon burning, the flame subsides. Further blowing will do more harm than good because the iron will begin to burn. An excessively high flame testifies to a violent burning of carbon; the bath boils up as in an open-hearth furnace and liquid metal may spurt out from the mouth. In this case the speed of the blower should be reduced and the amount of the blast supplied decreased by half. The air pressure should not exceed 800-1,000 mm w. c. The height of the flame becomes normal again.

After the end of the second period when the blast is cut off, the ladle is filled with lumps of 70-per cent ferromanganese, 5 kg per ton, and then 8 kg of 45-per cent ferrosilicon and aluminium.

To remove sulphur, the pig iron is usually treated with magnesium before introducing it into the converter. The slag is immediately removed so that it might not enter the converter where the sulphur may again contaminate the metal.

Nickel and copper are introduced into the converter together with pig iron, while chromium, vanadium, molybdenum and other alloying elements are added directly into the ladle. The additions dissolve rapidly since, before tapping, the steel has a sufficiently high temperature and an excess of heat. The slag is retained in the converter and then poured into the ladle to protect the metal from secondary oxidation.

Acceleration of the reactions of burning the pig-iron admixtures by oxygen constitutes a new trend in the development of Soviet metallurgy.

The first experiments were undertaken in 1933-1936 by N. Mozgovoi. In 1937-1940 the workers at the Moscow Institute of Steel continued these experiments with the aim of reducing somewhat the content of carbon, silicon and manganese in liquid pig iron in ladles of various capacities to improve the mechanical properties of pig-iron castings.

Experiments on obtaining steel with the aid of oxygen were performed in 1937-1944 under the guidance of I. Bardin. The oxygen was introduced into the metal with the help of silico-graphite tubes.

In 1944, I. Bobrov and coworkers utilised for the first time technically pure oxygen for blowing pig iron in the converter

with a charge capacity of 1.5 tons. In 1946, V. Kondakov experimented with a 1.5-ton converter into which oxygen was supplied through the bottom tuyeres. The problem of the endurance of an oxygen tuyere was solved by G. Ivantsov who used a water-cooled tuyere with a copper nozzle.

The application of oxygen for blowing pig iron in a converter made it possible to use the basic lining and the open-hearth pig. iron and dispense with haematite low-phosphorus pig iron,
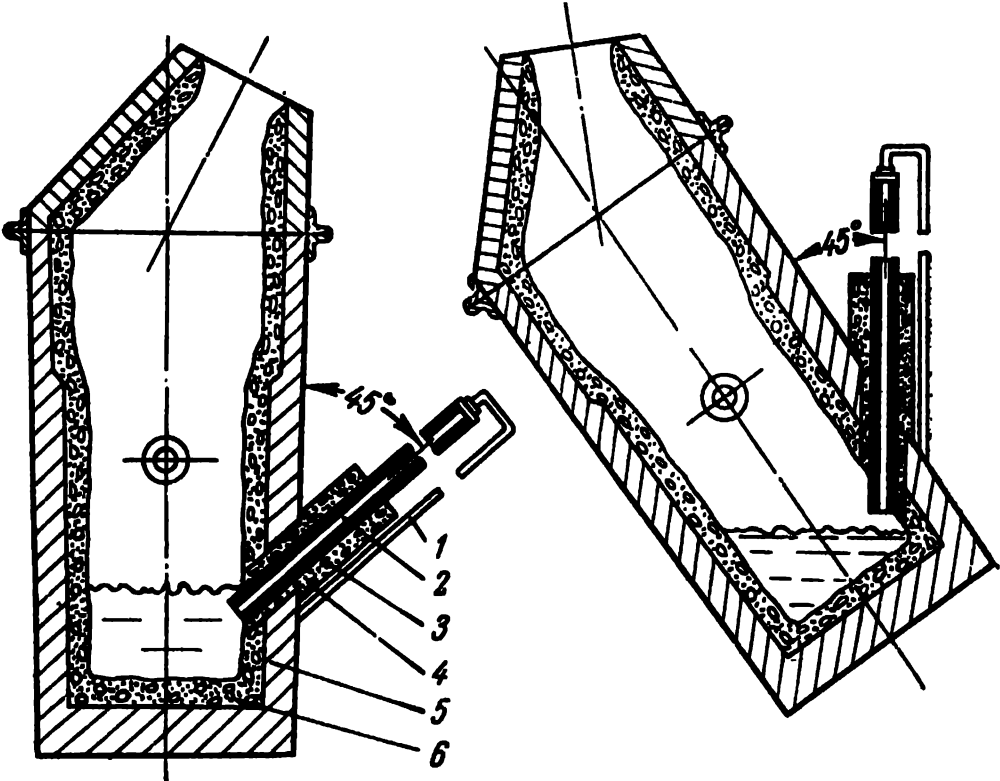

Fig. 140. Arrangement of oxygen tuyeres in a converter (Bobrov, Mozgovoi et al):

1—copper tube; *2*—quartz tube; *3*—graphite electrode; *4*—sand-clay lining; *5*—Dinas; *6*—layer of slag

the deficiency of which retarded the development of the side-blown converter process, and to obtain liquid steel with the same nitrogen content as in open-hearth steel.

According to V. Kondakov, the percentage of nitrogen in steel depends not only on its content in initial pig iron but also on the amount of oxygen in the blast (in per cent):

O_2 in blast	N_2 in steel	N_2 in pig iron
100	0.0021	—
100	0.0010	—
100	0.0013	0.0064
75	0.0022	0.0030
50	0.0069	—

Basic steel obtained by the side-blown converter process is by no means inferior in ductility and cold resistance to the basic open-hearth steel.

The tilt of the side oxygen tuyeres was increased to 45°. When the converter vessel was tilted during blowing to protect the tuyeres from burn-back their endurance increased (Fig. 140) since the tuyeres were raised above the metal.

From 1947 on, the oxygen side-blown converter process has been widely used at the Moscow plants "Dynamo", "Stankolit", etc.
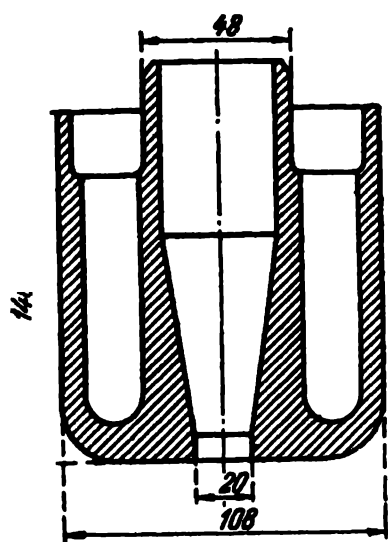
Fig. 141. The head of a water-cooled tuyere for blowing pig iron with technically pure oxygen (Novo-Tula Iron and Steel Works)

In Austria and the Federal Republic of Germany the blowing of pig iron with oxygen in converter bodies with basic linings was introduced in 1952 in the towns of Linz and Donawitz.

From December 1954, the foundrymen at the Novo-Tula Iron and Steel Works successfully employ oxygen fed through a water-cooled vertical tuyere (Fig. 141). The water is supplied at a pressure of 6-7 atm with a rate of 8-10 lit/sec. The tuyere is introduced through the mouth. The tuyere can endure over 100 heats. At first, 5 to 8 per cent lime is charged into the converter (depending on the silicon content in the pig iron), and then 1.5 tons of pig iron is poured in. The converter is placed in a vertical position and the blast is put on. The tuyere is lowered by an electric winch and the level by which the tuyere has been lowered into the converter is determined with the help of a special gauge.

The blowing is conducted at an oxygen pressure of 3-4 atm. The tuyere is 450-800 mm above the metal (from the level of quiet bath).

When the initial pig iron contains about 0.2 per cent P, no more than 0.016-0.045 per cent of it remains in the ready steel.

The best results were obtained by N. Mozgovoi and coworkers who used 2.5-3 per cent of fluor spar with 73-88 per cent CaF_2

and 7-14 per cent SiO$_2$. To reduce the spurting during violent boiling action the supply of oxygen was cut off and compressed air was fed instead. The data on five heats of steel with 0.5 per cent C are illustrated in Tables 48 and 49.

Table 48

Data of Material Balance

(Mozgovoi and coworkers)

Used	Amount, kg	Obtained	Amount, kg
Pig iron	100	Steel	87
Air	12.3	Slag	11.73
Oxygen	7.2	Spurts (metal and slag)	6.18
Lining	1.72	Gas	17.5
Lime	7.36	Blast losses	2.55
Fluor spar	2.35	Melting loss and smoke (F$_2$O$_3$) . . .	2.14
Total	130.93	Total	130.93

Table 49

Data of Heat Balance

(Mozgovoi and coworkers)

Input	Heat		Consumption	Heat	
	cal	%		cal	%
Heat from pig iron and blast . . .	24,727	46.7	Heat of steel	28,744	54.1
Carbon oxidation	12,678	23.82	Heat of slag	8,174	15.4
Silicon oxidation	7,540	14.23	Heat of gases	7,364	13.9
Oxidation of manganese, phosphorus and iron . .	8,139	15.3	Heat carried away with the smoke (Fe$_2$O$_3$,Fe$_3$O$_4$)	1,100	2.0
			Radiation and convection losses . . .	6,407	12.1
Total	53,084	100	Total . . .	53,084	100

The content of oxygen in the blast varied within 46-50 per cent at an average content of 0.98 per cent Si, 1.97 per cent Mn and 0.235 per cent P in the pig iron.

When the amount of sputtering was decreased from 3.8 per cent of the weight of the pig iron to zero, the yield of good steel increased from 87 to 90-92 per cent.

The side-blown converter process with the use of oxygen has every possibility for further development in steel foundries provided the cost of oxygen is reduced.

43. OPEN-HEARTH PROCESS

The steel for shaped castings is ordinarily made in open-hearth furnaces with a basic lining. The acid open-hearth process is used very seldom. As distinct from making steel for ingots the open-hearth process to produce shaped castings is conducted solely by the scrap-process method on a hard charge. The consumption of hard pig iron per heat amounts usually to about 30-35 per cent of the weight of the metal charge which is quite enough to reduce the phosphorus content from 0.09 to 0.03 per cent and to degas the bath during boiling. The rest of the metal is the returns from the home scrap or bought steel scrap.

The charge for open-hearth furnaces in foundries weighs about 10-50 tons, with the exception of heavy engineering plants where furnaces with capacities of 100 tons and more are constructed. For very large castings the metal may be tapped from several furnaces simultaneously. In such cases, the moulds are filled from several ladles, each of which is provided with two stoppers. For light castings the metal is poured into two ladles through a bifurcated spout.

The melting in an open-hearth furnace can be intensified either by enriching the blast with oxygen or, in the absence of the latter, with air compressed to 6-10 atm and introduced directly into the bath through an iron tube wound with wire and coated with a chrome-magnesite compound mixed with soluble glass. The roof of open-hearth furnaces is made of Dinas brick. At some plants the roofs are made from chrome-magnesite brick in which case the life of the roof can be increased approximately two times. A Dinas roof can endure not less than 350 heats and not infrequently 600 or 700 heats. The hearth of such furnaces is lined with chamotte brick overlayed by a course of chrome-magnesite brick, and above that two courses of magnesite brick. The hearth is lined with a fused layer of dead-burned magnesite or dolomite mixed with 15-25 per cent of open-hearth slag. Along the vertical section of the hearth brickwork, the insulation layer should account for 60-100 mm, the chamotte

brick for 100-300 mm, the magnesite or chrome-magnesite brick for 150-300 mm and the fused layer above for 150-300 mm.

The hearth should be burned in with the furnace strongly heated, in courses 6 mm thick after every 2.5 hours. The burning in of the hearth of a new furnace with a charge capacity of 30 tons takes not less than 6 days to complete. Dolomite takes a longer time to fuse on.

At one of the Soviet plants which produces Diesel engine locomotives the hearth is given a preventative maintenance after every 50 heats since after intensive wear it is impregnated with silicates and the vigorous boil of the bath is slowed down. This deteriorates the quality of the steel produced.

As a rule, open-hearth furnaces are fuelled by fuel oil. Only very few furnaces operate on a mixture of coke-oven and blast furnace gas or on natural gas. Some plants use generator gas as fuel.

In the case of the basic open-hearth scrap-process the charge includes open-hearth pig iron with 0.5-2 per cent Mn. When the content of manganese in the charge is low, the amount of lime should be increased to bind as much sulphur as possible in the slag. For example, with 5-6 per cent lime of the charge weight it should contain not less than 1 per cent Mn, but the quality of the cast steel will not deteriorate if there is only about 0.20 per cent manganese after melting. The amount of lime should be increased to 7-9 per cent.

The steel portion of the charge, especially gates, risers and rejected castings, are supplied for remelting after they are cleaned from burnt-on mixture which contains much silicic acid which tends to dissolve the basic lining. The metal chips are fed into the furnace in the form of briquettes or in compacted packages.

As a rule, melting is performed in basic open-hearth furnaces with the primary slag raked off after it has melted down.

When the melting is too soft, the melt is carburised by adding pig iron or ferromanganese. In the case of hard melting the period of oxidation is delayed too much which is uneconomical since the extra iron required will cool the bath.

The intensity of the standard open-hearth process with a hard charge depends on the time of separate periods of the furnace operation from tap to tap. The most advantageous is the method which ensures the most rapid completion of all operations and the production of well heated and well deoxidised steel with the required chemical analysis.

After tapping, the hearth, banks, brickwork between the ports, charging door sills and taphole should be carefully inspected. The remains of metal are removed from the holes in the hearth and new layers of magnesite or dolomite powder are fused

on. In this case the furnace should operate at its total heat capacity. It usually takes from 40 minutes to 1 hour to repair the hearth, banks and taphole.

The furnace is charged first with a thin layer of sheet scrap and iron ore. The second layer of light iron scrap can be introduced only after the first layer has been well heated through. The third layer is charged after the second layer has been heated through, etc. Heavy scrap is charged on top of the light one. The last layer consists of iron pigs which melt and flow down the lumps of heated steel and carburise it. Lime and iron ore are interlayed with the scrap and form the first slag which accumulates gradually first at the bottom and then rises to the surface as the metal becomes liquid.

The charging of hard burden takes from 2 to 3 hours. The duration of melting of these materials depends on the preheating temperature of the first layers. The higher is this temperature, the faster the metal melts and the less is the total time of the melting process.

It is good practice to increase the flame temperature during charging and melting by enriching the air with oxygen (up to 30 per cent) supplied at a pressure of 6-8 atm. Some 30-40 cu m of oxygen is consumed per ton of the metal.

The time of charging and melting takes about 75 per cent of the whole melting period and for this reason the various methods to intensify this process evolved today are aimed mainly at accelerating this process.

The use of a chrome-magnesite roof instead of a Dinas roof makes it possible to raise the temperature of the flame to 1750°C, instead of 1650°C with a Dinas roof.

Very much time is saved during charging and melting when cupolas are used to remelt steel scrap. Thus, at some plants in the Federal Republic of Germany the use of a cupola with a capacity of 9-10 tons of pig iron per hour with a basic lining and a blast raised to 550°C increased the output of two open-hearth furnaces of 35-ton capacity each by 40 per cent.

The cupola may be used to remelt either 100 per cent of steel scrap or 100 per cent of low-phosphorus pig iron (0.2-0.3 per cent P). The consumption of coke is 17-19 per cent, of limestone 5-8 per cent and of fluor spar 1-2.5 per cent. About 130 cu m of water per hour is needed to cool the cupola lining. With 100 per cent scrap in the charge the mean carbon content increases to 3.48 per cent and with 100 per cent of open-hearth pig iron it is increased to 4.8 per cent. With 25 per cent of open-hearth pig iron in the charge the sulphur content in the tapped pig iron decreases to 0.039 per cent and when using scrap alone to about 0.042 per cent. There is about 0.25 per cent silicon in pig iron. The use of liquid pig iron from the cupola makes it possible

to increase the weight of burden of open-hearth furnaces from 35 to 55 tons. A decrease in the carbon content to 0.5 per cent in the metal melted in a cupola is effected by blowing pig iron in the ladle. The consumption of oxygen is 30-40 cu m per ton of the metal.

That the pig iron can be blown in the ladle with oxygen was shown by Soviet investigators back in 1939.

The second period of melting—boiling—sets in after the charge has melted down and the slag has appeared. About 30-40 per cent of primary slag is raked off from the furnace after the charge has melted down to dephosphorise and decrease the thickness of the layer of nonheat-conductive slag. As the temperature of the metal and slag is raised the process of the carbon burning gradually grows in intensity and the bath begins to boil.

Boiling is a useful process both from the viewpoint of decarburisation and degassing of steel and obtaining a solid structure of the cast metal. It is a well-known fact that the partial pressure inside a bubble of a gaseous carbon monoxide rising to the surface will be negligibly small as compared with hydrogen and nitrogen dissolved in the metal. This creates favourable conditions for the evolution of other gases with the aid of carbon monoxide when the metal is being treated in a bath. This has been confirmed by the analyses of gases in tests taken during various periods of the melting operation. Degasification of the metal becomes the greatest towards the end of the oxidation period.

As we know, in an open-hearth furnace, carbon is oxidised by the oxygen present in the atmosphere of the melting chamber. In this case, the bath acquires 2.5×10^6 cal of heat per ton of the metal while when the oxidation is effected with the oxygen of the ore the bath loses 4.5×10^6 cal of heat per ton of the metal. An addition of 1 per cent of iron ore will decrease the bath temperature by about 40 per cent and 0.17 per cent of carbon may be oxidised.

By force of this fact, an addition of ore after hard melting, for example, excessively cools the metal. Every addition of the ore should never exceed 2 per cent of the weight of the metal. Hence the conclusion that when selecting a charge for an open-hearth furnace the excess of carbon should never exceed 0.5-0.6 per cent as against the carbon content specified for the final composition of the metal.

Today, we know several methods to intensify the process during the oxidation period of open-hearth process. According to one method, compressed air is fed through a water-cooled tuyere in the roof of the furnace. The end of the tuyere is lowered into the bath to a depth of 200 mm under the slag level. To

protect the tuyere from burn-up it is coated with a layer of chrome-magnesite compound.

Compressed air is blown through for only 10-15 min, at a rate of 10-15 cu m per min. The rate of the carbon burn-out is 0.05-0.10 per cent per minute or 3-6 per cent per hour, i. e., it increases several times.

The process of supplying oxygen directly to the bath of an open-hearth furnace during the oxidation period was studied in detail by I. Kazantsev, and the feeding techniques by the workers at the Central Research Institute of Iron and Steel Industry and the engineers of the "Zaporozhstal" Plant. The best method of supplying oxygen for blowing the metal is to feed it along two vertical water-cooled sprues through the furnace roof in order not to interfere with the work of steel-makers at the front wall of the furnace. The practical experience goes to show that the cost of blowing the metal with oxygen is higher than that with compressed air and the erection of compressors for the air is less expensive than the construction of an oxygen plant.

The principal aim of the open-hearth process control is to achieve a high temperature of the metal and slag at the end of the oxidation period. If the charge has been calculated for carbon content after melting correctly and the metal test has confirmed this, then the process of oxidation proceeds in a hot furnace with sufficient intensity and terminates within the shortest time. While there takes place the so-called "ore" boiling, i. e., decarburisation of the metal in the presence of iron ore in the slag, the metal gets gradually decarburised and heated. At the end of the operation we observe a "pure" boiling when the ore has been already melted and the metal is being oxidised by the oxygen present in the furnace atmosphere. During the "pure" boiling all metal is raised to the maximum temperature. No more ore is added, and attention is given to the state of the slag and its basicity. Preparations are underway for the last period—refining of the metal, the process of removing nonmetallic inclusions, and especially sulphur, from the molten metal.

Towards the end of the oxidation period the basicity of the slag is gradually increased by adding small portions of lime. The last, strongly basic, viscous oxidising slag covering is formed with the length of the run in the viscosimeter of 40-50 mm.

If the melting process has been very hot and the metal and slag are raised to the highest safe temperature, the blast and the amount of fuel are decreased. Manganese is reduced from the slag and passes to the metal and the sulphur passes from the metal to the slag, as the oxidising ability of the slag drops.

After the metal has been tested for carbon and sulphur, depending on the grade of steel being melted, there sets in the final

melting period. Prior to deoxidation, the composition of the slag should be approximately as follows: 45-50 per cent CaO; 15-20 per cent SiO_2; 6-10 per cent FeO; 2-4 per cent Fe_2O_3; 8-12 per cent MnO; 3-5 per cent MgO; 1-2 per cent Al_2O_3. Its basicity is 2.3-2.8. The total duration of the final melting period should take about 40-50 min.

The final melting starts with predeoxidation of the bath by ferromanganese (to obtain about 0.5-0.8 per cent Mn in the metal) and adding, 8-10 minutes later, red-hot blast-furnace ferrosilicon (0.10-0.17 per cent Si). The main aim of predeoxidation is not only to reduce the iron oxides in the metal and the slag but also to enable the formation of fusible manganese silicates which will merge into larger particles and dissolve in the slag, i. e., the final aim is to deoxidise and refine the metal.

Fig. 142. Change in the coefficient of sulphur distribution between the slag and metal depending on the ferrous oxide content in the slag (Umrikhin)
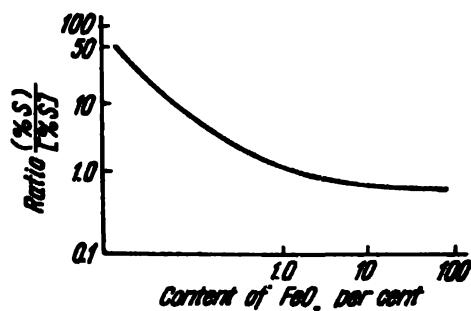
Desulphurisation of steel is extremely important for obtaining sound steel castings.

At low concentrations of carbon, the equilibrium content of sulphur can be found from the equation

$$[\% \ C][\% \ S] = 0.11. \qquad (203)$$

Since the carbon content is connected with the equilibrium content of oxygen in the metal

$$[\% \ C][\% \ O] = 0.0025 \qquad (204)$$

we shall obtain with the given carbon content from the comparison of both equations the equilibrium relationship between the sulphur and oxygen content in the metal

$$[\% \ S] = 4.4 \ [\% \ O]. \qquad (205)$$

In this way, the operations of deoxidation and desulphurisation of the metal proceed simultaneously and, apparently, the better is the metal deoxidised, the more sulphur will be eliminated from it.

The direct relationship between the content of FeO in the slag and the coefficient of distribution of the sulphur between the slag and metal (%S)/[%S] is expressed by the curve in Fig. 142.

In turn, the coefficient $\frac{(\% \text{ S})}{[\% \text{ S}]}$ can be found from the equation:

$$\frac{(\% \text{ S})}{[\% \text{ S}]} = N_s \left\{ 2.6 + 1.6 \frac{(\text{CaO}) - 4\,(\text{P}_2\text{O}_5)}{(\text{SiO}_2)}\,(L' - 2) + 11\,[\% \text{ Mn}] \right\}, \quad (206)$$

where N_s = total number of moles in 100 g of the slag;

$$L' - \frac{(\text{CaO}) - 4\,(\text{P}_2\text{O}_5)}{(\text{SiO}_2)} = \text{basicity of the slag.}$$

This equation shows the positive role of manganese introduced into the metal during predeoxidation and also during the process of reducing MnO from the slag which can sometimes be achieved by adding manganese ore to the slag during the process of pure boiling.

These conclusions on the effect of the manganese during the basic open-hearth melting process tell us that it is not obligatory to maintain a high manganese content in the metal during the initial phase of the melting. In other words, air refining may be performed at a low manganese content after fusion and during the ore boiling. But the positive effect of manganese during desulphurisation of the metal at the end of hot melting cannot be denied.

The relationship between the oxygen and sulphur content in the metal high in carbon (1.2-1.7 per cent) can be expressed by Eq. (207), as in Eq. (205),

$$[\% \text{ S}] = 3\,[\% \text{ O}]. \quad (207)$$

Thus, even a partial skimming of the primary slag after fusion can be regarded as a substantial preparation for the subsequent removal both of phosphorus and sulphur. The useless silicic acid goes together with the slag, and the basicity of the slag is therefore increased at the end of the melting.

The investigation of test heats in open-hearth furnaces enabled P. Umrikhin to draw the following conclusions:

1. The process of desulphurisation of steel in a basic open-hearth furnace can be commenced when the charge is being melted, and the slag can be removed at high coefficients of distribution. A good covering of basic slag formed early in the heat facilitates this operation and enables steel low in sulphur to be obtained.

2. The process of desulphurisation during the final melting is fully dependent on the intensity of the bath boiling, on the basicity and diffusion ability of the slag.

3. Reduction of manganese from slag during melting tends to increase the rate at which sulphur passes from the metal to the slag.

4. The temperature factor is of certain importance in the kinetics of desulphurisation of the metal; however, the superheat

of the metal alone cannot speed up the process of sulphur removal.

The deoxidation of steel—removal of oxygen from the metal—comprises two operations: the first provides for the transfer of the oxygen of ferrous oxide into the products of deoxidation reactions and the second—for the transfer of dispersed oxides into larger agglomerates and their transfer to the slag. The investigations conducted during the last five years have shown that the nature of oxide inclusions in the ready steel and, consequently, the change in the structural and mechanical properties of steel depend on the sequence of introducing the deoxidisers.

The amount of nonmetallic inclusions, on which the ductile properties of steel depend directly, is predetermined by the method accepted for the predeoxidation of steel.

The minimum amount of nonmetallic inclusions was obtained when the steel was predeoxidised by silicomanganese with the manganese-to-silicon ratio in this alloy equal to 6 or 7, and not first with ferromanganese and then with ferrosilicon, or conversely. The shape, size and distribution of nonmetallic inclusions are determined by the nature of the products of the final deoxidation reactions.

Not less than 15-20 minutes are required to dissolve the silicomanganese completely in a steel bath and remove the products of the deoxidation reactions. The final mixing and the completion of the refining process take place when the metal is tapped into the ladle and at a corresponding holding of the steel in the ladle.

The duration of final melting after preliminary deoxidation depends on the given grade of steel. Alloying of steel is completed by adding corresponding ferroalloys in the order of their affinity for oxygen. The smaller is the affinity of the alloying admixture for oxygen, the less admixture will burn in the furnace and the less will be the melting losses. During remelting, copper, nickel and molybdenum may pass from the scrap almost completely and for this reason additional quantities of them may be introduced before the other admixtures, for example at the start of the reduction period.

After predeoxidation, ferrochromium raised to red heat is introduced in the form of large lumps. The admixtures with a small specific gravity, for example, aluminium, graphite, anthracite or charcoal, are added to the stream on the spout. The latter three are supplied in paper sacks. About 1-1.2 kg of coal or graphite is required per ton of the metal to increase carbon content in steel by 0.01 per cent.

Vanadium, titanium and other elements having a tendency towards a strong oxidation must be introduced only after the bath has been deoxidised finally with 45- or 75-per cent

ferrosilicon and aluminium. They are usually added into the ladle.

Alloying of liquid steel is the subject of detailed study in the course of steel metallurgy.

The steel is deoxidised finally under conditions in which the minimum amount of oxygen is left in the metal for the last deoxidisers. This is achieved by adding into the deoxidised steel aluminium as well as titanium, zirconium, vanadium, niobium, boron and other elements which have the greatest affinity for oxygen.

Aluminium is the strongest deoxidiser. At 1600°C its constant is

$$K = [\% \ Al]^2 \ [\% \ O]^3 = 1.35 \times 10^{-14}. \qquad (208)$$

However, deoxidation with aluminium may cause various undesirable phenomena in the structure of the castings and especially of their heavy sections. For example, the castings and ingots develop streaks filled with liquids (sulphides and silicates) and, in the absence of sulphur, centreline porosity and, apparently, intercristalline pores (Fig. 143).
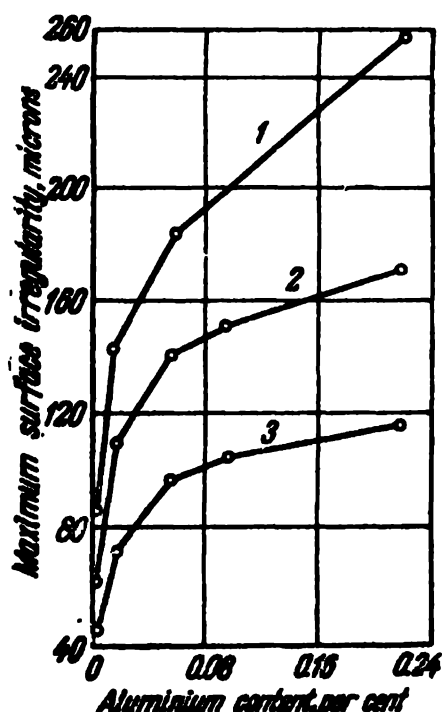
Fig. 143. Change in the irregularities on the surface of transverse templates depending on additions of aluminium (ingot diameter 90 mm) (Kasum-zade):

1—zone of columnar crystals; 2—intermediate zone; 3—zone of globular crystals

A. Baikov was against deoxidation of steel with aluminium and believed it quite unfit for the purpose.

The common viewpoint held for a long time as to the utilisation of aluminium as steel deoxidiser is today rather disputable.

However, when added in amounts of 0.05 per cent aluminium is known to prevent the development of underskin cavities in steel castings.

In our personal practice we obtained good ductile properties of cast carbon steel 20ЛŽ-30Л with additions of 0.06 per cent Al at an increased silicon content in the ready metal (0.5-0.6 per cent) as well as from the use of alumosilicomanganese alloy AMS (10 per cent Al, 18 per cent Mn and 14 per cent Si).

The aftereffects of deoxidation of steel with aluminium are now being studied at a number of heavy engineering plants, and no final conclusions have been made so far.

There are also methods for a diffusion deoxidation of steel with a compound consisting of 40 per cent of slaked lime and 60 per cent of aluminium powder introduced into the slag and also with a synthetic slag prepared in another furnace. Although the latter method gives certain additional advantages, it is not so widespread because of its high cost. On the other hand, deep desulphurisation of steel in the ladle becomes quite practicable.

The melting of a high-alloy steel in a basic open-hearth furnace with the charge capacity of 10 tons may be illustrated by the melting procedure adopted at the "Serp i Molot" Iron and Steel Works in Moscow for grade Г13Л steel.

The charge is prepared from open-hearth pig iron and steel scrap with a view to obtaining 0.45 per cent C upon melting. The phosphorus content in the pig iron should not exceed 0.12 per cent.

The principal aim of the operation is to obtain a metal free from manganese oxides which are difficult to remove from the bath and which deteriorate the resistance of the steel to cracking. The melting process is divided into two periods. During the first period every effort is made to obtain a completely deoxidised mild metal with 0.10-0.12 per cent C while the second period is accompanied by alloying the metal with manganese with good care taken to prevent it from oxidation.

The primary slag should be always skimmed if the phosphorus content in the metal exceeds 0.06 per cent and the slag is rich in magnesia.

The boiling action is maintained by adding iron ore. The rate of carbon burning during the period of pure boiling should be 0.20-0.66 per cent an hour; it is adjusted by feeding some lime. The slag basicity is 2.20-2.60. After the carbon has burned out to 0.15 per cent, the metal is deoxidised with ferromanganese with a view to obtaining in the metal not less than 0.4 per cent Mn and then a preheated 12-per cent ferrosilicon is added in amounts of 10-12 kg per ton.

Before the bulk of the ferromanganese is charged into the furnace, 1.5-2 kg of ground oil coke per ton is added onto the slag.

The ferromanganese is heated on a shovel and introduced into the furnace in amounts of 260 kg per ton of liquid steel.

The final deoxidation is done with slaked lime and aluminium powder—20 kg per ton. The supply of fuel oil into the furnace is cut off and the smoke gate is closed. The slag should contain not less than 40 per cent CaO and not over 12 per cent MnO with the basicity 2.6-3.2.

The final melting by silicon is conducted by adding into the spout 5-6 kg of 45-per cent ferrosilicon and 0.8-1 kg of ground aluminium per ton of the metal.

The deoxidation of the slag due to the additions of slaked lime and aluminium powder is illustrated in Table 50.

Deoxidation of Slag Due to Additions of Slaked Lime and Aluminium Powder

Process	Number of heats	Mean content, %, of				Mean basicity of the slag
		CaO	MnO	FeO	SIO$_2$	
Without lime and Al powder	58	34.34	20.14	6.72	22.61	1.51
With lime and Al powder	54	45.20	7.70	2.42	17.00	2.66

N. Dobrokhotov recommends that open-hearth steel should be tapped into the ladle in a boiling state. Deoxidation and alloying should be performed with cold ferroalloys not in the furnace but only in the ladle (in melting carbon and low-alloy steel). Aluminium in amounts of 2 kg per ton of the metal should be added to the ladle for deoxidation.

44. ELECTRIC ARC PROCESS

An electric arc furnace is a simple and convenient steel-making unit intended for small and medium castings. They are widely used at Soviet plants, especially at automobile and tractor works as well as at heavy engineering plants. The charge capacity of the furnaces varies from 0.5 to 10 tons. Heavy engineering plants use furnaces with capacities of up to 30 tons for large castings made from alloy steel and 80-ton electric furnaces, and larger are being designed for new plants.

The atmosphere of an electric arc furnace is deficient in oxygen and has a reduction character which makes it possible to reduce the content of FeO in the slag to 1 per cent at the end of the heat and even below that and to tap the metal free from nonmetallic

inclusions and alloy the steel with any element at the minimum melting loss.

The melting in an electric furnace can be carried out by one of the following four methods:

(1) in a basic furnace, oxidising the admixtures with iron ore;

(2) in a basic furnace without additions of iron ore (fusion of steel scrap);

(3) in an acid furnace with oxidation;

(4) in an acid furnace without oxidation.

The choice of these methods depends on the grade of steel, its sulphur and phosphorus content and the charge materials available containing various amounts of sulphur and phosphorus.

Furnaces with a basic lining are employed for critical castings made from high-strength structural alloy steel or from heat-resistant steel. This makes it possible to mix the bath during boiling, degas the metal, remove a greater part of the phosphorus and a part of sulphur with the primary slag, deoxidise and desulphurise the steel at the end of the heat and obtain the content of almost all admixtures within the narrow specified limits.

If there is much unoxidised steel scrap in the foundry, such as cuttings from the forging and rolling departments or the home scrap, then the melting may be performed in a basic furnace without oxidation if the same requirements are made to castings from alloy steel. This makes it possible to increase the output of the furnace.

Melting in acid electric arc furnaces must be conducted only on pure scrap. Oxidised and light-weight scrap is remelted with an addition of 5-10 per cent of pig iron and the steel is therefore produced with oxidation and boiling while unoxidised scrap free from sulphur and phosphorus is remelted without oxidation. The latter method possesses numerous technical and economic advantages and is extensively used at engineering plants.

Special investigations were undertaken in the Soviet Union to compare the properties of steels obtained in electric arc furnaces with a charge capacity of 3 tons with a basic and acid linings. The data are illustrated in Table 51.

The basic melting process was conducted under the following conditions: charge composition—20-30 per cent of small scrap and chips; 30-40 per cent of gates and risers 5-20 kg in weight; 30-40 per cent of large risers and scrap 20-60 kg in weight; 10 per cent of pig iron.

After the steel was tapped the furnace was thoroughly inspected, the banks were burned in and the holes filled. Chips and small scrap were charged onto the hearth, then large scrap, iron pigs and above that again small scrap and chips. After charging the electrodes were built up.

Table 51

Comparative Data of Electric.Arc Furnaces

(Kryanin and Sushko)

	Plant A	Plant B
Electric furnace:		
type	ЭРУ	ДСТ
lining	basic	acid
shell diameter, mm	2,600	2,090
inner diameter of bath, mm	1,900	1,990
depth of bath, mm	360	450
distance from hearth to roof, mm	1,250	1,450
Transformer:		
power, kW	1,500	1,500
Voltage, V:		
primary	6,000	6,000
secondary	175-123	210-121
rated current, A	144	144

The melting was carried out intensively, with a 20-40 per cent of overload on the transformer. The hard remains of the charge were pushed from the banks into the liquid steel. At the end of the heat two or three shovelfuls of powdered ore and fluor spar were thrown into the furnace. The increase in carbon was 0.40 per cent as compared with the final composition of the steel.

Upon melting, all the primary slag was raked off and new slag was formed. The skimmed slag was composed of 35-53 per cent $CaO+MgO$; 15-30 per cent FeO; 4-8 per cent MnO; 14-33 per cent $Al_2O_3+SiO_2$.

The oxidation of the elements per minute was on an average 0.0093 per cent for carbon, 0.0052 per cent for manganese and 0.0006 per cent for phosphorus.

After the oxidising slag had been removed the furnace was charged with 4-8 kg of ferromanganese per ton of the metal and "white" slag was formed by introducing 50 kg of lime and 10 kg of fluor spar. The transformer was switched over to the "star" circuit to reduce the voltage. After the slag mixture had melted down the furnace was charged with a mixture consisting of coke (one part), lumps of lime (ten parts) and fluor spar (one part) for a diffusion deoxidation of the slag. In the molten mixture the slag contained 65-72 per cent $CaO+MgO$; 1-3 per cent FeO; 1-3 per cent MnO and 22-28 per cent $SiO_2+Al_2O_3$. After that, a mixture of ground coke, lime, 75-per cent ferrosilicon and fluor

spar was added to the slag at a ratio of 1:10:3:1. The slag acquired a white-to-gray colour and crumbled to powder on cooling.

The steel was held under this slag for 1 hour. After testing the metal for temperature and deoxidation, 4-5 kg of ferrosilicon was introduced into the bath per ton of the metal as well as ferromanganese in conformity with the analysis of the ready steel. The temperature of the metal was 1560-1570°C. At the end of the heat the slag contained 53-61 per cent CaO; 8-15 per cent MgO; 0.4-1.7 per cent FeO; 0.2-1.1 per cent MnO; 19-23 per cent SiO_2 and 2-5 per cent Al_2O_3. After that 1 kg of aluminium per ton of the liquid steel was introduced with a rod into the furnace and the metal was poured into the ladle. The steel was poured into the moulds at 1485-1495°C. The consumption of electric power was 825-1077 kW-hr per ton of steel.

The acid melting process was conducted in the following conditions. After cleaning the furnace hearth from the remaining slag it was burned in with a mixture composed of one part of the moulding mixture and one part of quartz sand. The hearth was charged with fine scrap (15-20 per cent) then with large-size scrap (50-70 per cent) and above that the fines (10-20 per cent). Then the current was switched on to the full transformer power until all the charge was melted down. Some sand was added to the furnace to produce a thin layer of slag. After a weak boiling for 10-30 minutes the transformer was switched over to the "star" circuit. The bath was stirred, with the current switched off, and a metal sample was taken. If the carbon content in the metal was close to the specified value the boiling was continued for some time after adding some iron ore to account for the carburisation of the metal due to the addition of ferromanganese. If there were less carbon than specified, pig iron was added to the bath or else the electrodes were submerged into the metal.

During the boiling period the slag contained an average of 40-55 per cent SiO_2, 25-30 per cent FeO and 10-15 per cent MnO.

The boiling was stopped by adding 45-per cent ferrosilicon in amounts of 4-6 kg per ton of the steel. The bath was stirred. Then the slag was thinned with 5-10 kg of lime per ton of the steel. Before tapping, the slag contained an average of 60-65 per cent SiO_2; 8-10 per cent CaO; 2-3 per cent Al_2O_3; 4-8 per cent FeO; 5-9 per cent Fe_2O_3; 8-10 per cent MnO.

After complete killing, 12-16 kg of ferromanganese per ton of metal was added to the steel. The metal was tested for deoxidation and temperature, and after receiving aluminium in amounts of 1 kg per ton, the steel was poured into the ladle.

The total electric power consumption per ton of the metal was 700-800 kW-hr. The steel was poured at 1550-1600°C after holding it in the ladle for 5-10 minutes.

The charge capacity of the furnace with a basic lining was 4,000 kg (with 33-per cent overload) 'and with an acid lining 3,000 kg.

The duration of the melting process and the output of the furnace are represented in Table 52.

Table 52

Duration of the Melting Process and the Output of the Furnace

Proc-ess	Duration, hr-min				Electric power consumption, kW-hr				Liquid steel melted, tons	Output of furnace per hour, kg	Specific consumption of electric power, kW-hr
	melting	boiling	refining	total	melting	boiling	refining	total			
Basic	1-14	0-46	1-16	3-16	1;660	876	1,121	3,657	3,850	1,191	950
Acid	1-16	0-17	0-30	2-03	1,385	297	345	2,027	2,750	1,355	738

Thus, an acid furnace has a considerable reserve of output (30-35 per cent). But even if the reserve is not utilised completely, the melting process in an acid furnace proceeds, all other conditions being equal, at a faster rate and less current is consumed (about 800 kW-hr/ton instead of 900 kW-hr/ton). The saving in the electric power amounted to about 22 per cent during a prolonged period of observation.

The life of the hearth in a basic furnace was on an average 2,900 heats and that of an acid one 10,900 heats. The endurance of the furnace walls was, respectively, 2,000 and 10,900 heats, that of the roof 150 and 400 heats. In the basic process 16 kg of electrodes were used per ton of steel and 14 kg in the acid process.

When comparing the costs of the materials, electric power and labour per ton of liquid steel the cost of one ton of acid steel proved to be 9 per cent less than the cost of basic steel. The annual saving in the refractories was 306 tons and in the electric power 1.5 million kW-hr when a 3-ton furnace was changed over from the basic to the acid process.

The statistical processing of hundreds of basic and acid heats has yielded average figures of the mechanical properties of carbon steel grade 30Л most widespread in the foundry practice (Table 53).

The mean chemical analysis of the steel from the investigated basic heats was as follows: 0.31 per cent C; 0.61 per cent Mn; 0.27 per cent Si; 0.02 per cent P and 0.018 per cent S, and from acid heats: 0.31 per cent C; 0.80 per cent Mn; 0.34 per cent Si; 0.049 per cent P; 0.043 per cent S. It can be easily seen that the acid steel is stronger but less ductile while the basic steel is physically

Table 53

Comparative Mechanical Properties of Grade 30Л Steel Made in a Furnace with a Basic and Acid Lining

(Kryanin and Sushko)

Heats	σ_s, kg/mm²	σ_b, kg/mm²	δ_s, %	ψ, %.	a_k, kgm/cm²	*Bhn*
Basic process						
Ordinary	27.2-31.3	48.1-54.3	28.6-38.8	39.7-47.4	—	—
Investigated	28.5-35.0	47.0-65.5	27.0-37.0	24.5-47.5	5.9-9.8	143-170
Average of investigated heats	30.3	53.7	31.7	41.4	7.6	158
Acid process						
Ordinary	31.3-34.0	52.5-57.7	16.9-22.5	26.3-37.1	—	—
Investigated	31.8-38.2	57.3-70.7	14.0-20.6	17.2-29.6	3.9-5.0	156-207
Average of investigated heats	35.4	62.2	16.1	20.0	4.2	178

more homogeneous than the acid steel with a higher content of phosphorus and sulphur.

Comparison of the content of gases in the steels of the investigated heats provides an estimation of the basic and acid processes (Table 54).

Table 54

Content of Gases in the Basic and Acid Processes in an Electric Furnace, per cent

(Kryanin and Sushko)

Gas	Basic process				Acid process			
	with charge melted down	during steel boiling	before tapping of steel	in pouring into moulds	with charge melted down	during refining	before tapping of steel	in pouring into moulds
Oxygen . .	0.0355	0.0411	0.0121	0.0121	0.0294	0.0140	0.0153	0.0090
Nitrogen . .	0.0070	0.0102	0.0071	0.0071	0.0071	0.0072	0.0065	0.0049
Hydrogen . .	0.00042	0.00065	0.00031	0.00075	0.0008	0.0003	0.00023	—

When poured into the mould acid steel showed a lower oxygen content than basic steel. It can be assumed that at a high sulphur content in acid steel (0.043 per cent) as compared to basic steel (0.018 per cent S) its oxygen content is also high [%S]= =4.77 [%O].

When the sulphur content is correspondingly reduced in acid metal there is a possibility of decreasing its oxygen content and, consequently, sharply increasing the ductility of the steel due to the smaller amount of harmful admixtures—oxysulphides.

No decrease in the nitrogen content was registered in the basic process from the start to the end of the melting while during the acid process its concentration decreased by 2/3 of its original amount. In this way, acid steel has a marked advantage as regards its nitrogen content.

Under comparatively similar conditions the fluidity of basic steel was somewhat poorer than that of acid steel.

The following conclusions can be made from the comparison of the basic and acid processes. The acid process offers undoubted economic advantages when castings of ordinary quality are produced. But when the castings are to have better ductility and, consequently, better physical homogeneity preference should be given to the basic process.

The principal shortcoming of the acid process is that phosphorus and sulphur cannot be transferred to slag.

When using the same grades of carbon or low-alloy steel made in several furnaces, as at tractor works, for example, it will be good practice to line a part of the furnaces with basic materials and conduct the process on the wastes of acid steel with a low nitrogen content and with much sulphur and phosphorus, and line the other furnaces with acid materials and use them for re-melting the wastes from the basic steel which has little sulphur and phosphorus but much nitrogen. In any case the principal demand is for steel scrap selected not only by the amount of sulphur and phosphorus but also by its content of oxygen (degree of oxidation) and nitrogen.

The highest technical and economic level of foundry production can be attained if the home wastes are sorted out rationally, especially the wastes of alloy steel, and the method of melting steel in electric arc furnaces is selected correctly.

The utilisation of the wastes of chromium and other grades of alloyed structural and heat-resistant steels is one of the most complex problems which face metallurgists. According to an appropriate Soviet standard, the chromium content should not exceed 0.3 per cent in carbon steel heats and, therefore, the scrap and chips with an increased content of chromium are highly undesirable. The oxides of chromium which pass to the slag make it much thicker which slows down the burning of carbon and

phosphorus as well as the deoxidation and desulphurisation. It follows from this that chromium must be got rid of early in the heat by skimming the oxidising slag.

When melting chromium-containing grades of steel, the chromium should be preserved and as much of it as possible should be transferred from the wastes to the final metal, reducing the consumption of ferrochromium at the same time.

Chromium oxidises after silicon and manganese and for this reason when oxidising by one method or another the metal obtained from the chromium wastes any small quantity of chromium in the metal can be obtained at the end of the oxidation period of the heat.

The reducing atmosphere of an electric arc furnace and hence the deficiency of oxygen during the oxidation period of the heat involve the necessity of using iron ore which requires much electric power to be heated. Calcined iron ore is introduced into the furnace upon melting in amounts of 1 per cent, and 10 minutes later 60-70 per cent of slag is skimmed without switching on the current. Most of the phosphorus is removed from the furnace with the primary slag. Then the furnace is charged with 1-1.5 per cent of freshly burned lime and the metal is heated to 1540°C. Each 0.1 per cent of the carbon to be removed requires an addition of about 1 kg of ore per ton of the metal. The ore is introduced in small portions every 10 minutes after the boiling action begins to subside. The current is switched off, the bath is stirred, some metal is sampled and the course of the further operation is controlled by the express-analysis for carbon, manganese, phosphorus and chromium.

To make important grades of steel about 80 per cent of the slag of the oxidation period is skimmed for the second time and the "pure" boiling is continued, this time without introducing any of the additions.

This course of the oxidation process can be termed classical. The consumption of electric power increases. For this reason, and depending on the scale of production, new methods have been evolved to improve the electric arc process by using oxygen or preliminary blowing of pig iron in a converter, i. e., triplex-process cupola+oxygen converter+electric furnace.

Oxygen is fed into the furnace through a specially provided side port with the aid of a coated iron tube. It is good practice to cut the metal scrap with oxygen flame when the charge is being melted. A shield with a mica observation hole is provided to protect the steelmaker from sputtering. Iron ore may be supplied instead of oxygen in which case the bath will be rapidly heated up to such temperature (1800-1900°C) that the sequence of burning of the admixtures will change—carbon will be the first to burn and then silicon, manganese and chromium. This makes it

possible to conduct the melting with a large amount of chromium oxides in the slag. The slag no longer hampers the oxidation processes because the oxygen enters the metal bypassing the slag and, besides, the slag is mixed with the metal as a result of which the contact surfaces of the interacting phases increase.

N. Dobrokhotov recommends the following procedure for the final melting and deoxidation of carbon and low-alloy steels for shaped castings in basic electric arc furnaces:

1. A strongly oxidised and strongly basic slag should be formed and deoxidised with ground ferrosilicon and aluminium without the use of fine coke and reducing the content of ferrous oxide in the slag below 0.5 per cent to obtain white slag.

2. The metal should be kept under the white slag for a short time, after which first the slag should be tapped into the ladle and then the metal to mix them better and remove the maximum amount of sulphur from the metal into the slag.

3. The steel should be finally deoxidised and alloyed with cold ferroalloys only in the ladle and not in the furnace.

· The temperature of the metal being tapped should exceed the liquidus temperature by 100-150° depending on the pouring conditions.

4. The lime used to form the reducing slag should contain the minimum of moisture and sulphur.

5. If the content of sulphur in ready steel is to be reduced to 0.025 per cent, the metal should be tapped in a boiling state and the steel deoxidised and alloyed only in the ladle.

Proper utilisation of the wastes is very important in the production of stainless steel of grade 1X18H9T. The charge is composed of 70-100 per cent of the wastes of high-chromium and silicon steel so as to obtain the following composition after meltdown: 0.2 per cent C, 12 per cent Cr and 1.0 per cent Si. If the silicon content is too low the charge is enriched with 45-per cent ferrosilicon to prevent the chromium from burning. The more carbon there remains in the bath after blowing with oxygen, the less chromium will be burned. With 0.10 per cent C, the yield of chromium is about 90 per cent and with 0.06 per cent C—80 per cent.

The oxygen is introduced with the aid of iron tubes (3/4 inches) coated with a refractory compound. The gas is supplied at a pressure of 7-10 atm at the rate of 6-8 cu m per ton of metal.

After blowing the slag contains up to 40 per cent Cr_2O_3. To extract the chromium from the slag it is treated with ferrosilicon with 50 per cent Si, silicocalcium or silicoaluminium. When using silicochromium with 0.02-0.03 per cent C the expensive low-carbon ferrochromium may be dispensed with and the yield of chromium increased to 93 per cent. To increase the

degree of reduction of the chromium from the slag it is necessary to decrease its temperature which is achieved at the Kuznetsk Iron and Steel Plant by adding the wastes of stainless steel into the bath after blowing.

After deoxidation the slag is raked off, some metallic manganese and titanium are added and the metal is poured into the ladle.

Heat-resistant steel of grade X15H15K3 goes to make castings for steam and gas turbines. The valuable wastes of these steel can be fully utilised after blowing with oxygen the bath of the basic electric arc furnace with a charge capacity of 3-10 tons after melt-down. The steel is deoxidised finally by adding 0.01 per cent B and 0.3 per cent Al, or 0.30 per cent Ca and 0.10 per cent Zr. By its mechanical properties such steel is as good as the steel obtained from "fresh" materials without blowing with oxygen and is 30 to 40 per cent less expensive besides.

The following experimental data (Table 55) show the advantages derived from the blowing of carbon and low-alloy steels with oxygen.

Table 55

Operating Indices of Furnaces When the Metal is Melted with Oxygen and without Oxygen in a Basic Electric Arc Furnace with the Charge Capacity of 3 Tons

(Sokolov)

Steel	Melting method	Number of heats	Melting		Oxidation		Reduction		All heat	
			*	**	*	**	*	**	*	**.
Carbon	Without oxy-gen	65	1-52	496	0-37	160	0-41	154	3-10	810
	With oxygen	24	1-37	424	0-34	154	0-40	144	2-51	722
Low-al-loy	Without oxy-gen	24	1-47	482	0-45	192	0-42	141	3-14	815
	With oxygen	5	1-36	.442	0-37	159	0-36	126	2-49	727

* Duration, hr-min.
** Electric power consumption, kW-hr.

Coreless induction furnaces of high and low frequencies can be expediently employed to make light castings from high-alloy multicomponent grades of structural steel as well as from heat-resistant, stainless and corrosion-proof steels and cobalt, titanium, and zirconium base alloys.

Such furnaces have the following advantages:

1. The liquid alloy can be raised to the temperatures which cannot be attained in furnaces of other designs. The highest limit of the temperatures is determined depending on the refractoriness and endurance of the crucibles used for melting the alloy.

2. Comparatively small heat and electric losses since the highest temperature caused by induction currents is concentrated inside the bath.

3. The temperature and the melting rate can be adjusted at will.

4. The volumetric mixing of the bath under current speeds up the dissolution of hard additions of alloying admixtures and the reactions that occur in the metal and slag.

5. The minimum melting loss of valuable alloying admixtures allows a satisfactory utilisation of home scrap returns.

6. Short heat cycles, especially useful in foundry practice.

7. Possibility of producing thin-walled castings which require a higher superheat of the alloy during the filling of the moulds.

8. Possibility of obtaining a highly pure alloy free from gaseous and nonmetallic inclusions since an induction furnace can be easily isolated from the atmosphere and the melting conducted in the atmosphere of the gas or in a vacuum.

Among the shortcomings inherent to induction furnaces are complicated electrotechnical equipment and poor endurance of the crucible lining which cracks rapidly due to alternate expansion and compression.

These shortcomings limit not only the capacities of the crucibles but also the sphere and scale of the entire foundry production.

Coreless induction furnaces are very promising for melting expensive alloys since the saved valuable alloying components more than compensate for the production and capital outlays.

Today, the charge capacity of coreless furnaces employed in foundry shops varies from 45 kg to 8 tons and more. Laboratory furnaces have capacities up to 50 tons.

Table 56 illustrates the operating indices of coreless furnaces of various capacities.

The figures in Table 56 show that the insufficient power of the generator slows down the melting process and increases electric energy consumption per ton of steel since the electric and heat losses are proportional to the melting time. The greater is the furnace capacity, the smaller is the specific consumption of electric power. It stands to reason that when the process is continuous, less current is needed than in the case of large intervals in operation. The idle time of the furnace needed for repairs and charging the crucible should be reduced to the minimum. The greater the crucible endurance, the more economical is the furnace operation.

Table 56

Operating Indices of Coreless Furnaces of Various Capacities
(Edneral)

Furnace capacities, kg	Generator power, kW	Duration of heat, hr-min	Electric power consumption, kW-hr/ton	Output per hour, kg
45	60	0-40	850	70
180	100	1-00	800	130
270	150	1-15	750	225
450	300	1-00	700	450
900	600	1-00	675	900
1,800	600	2-15	700	800
1,800	1,200	1-00	650	1,800
3,600	1,200	2-15	675	1,575
5,400	1,750	2-30	675	2,200

The long life of the crucibles is one of the most decisive factors for the economical operation of coreless induction furnaces. As a rule, crucibles are made from a wet moulding mixture charged into the furnace. The compacted compound is carefully dried. Then, a hollow iron cylinder with perforations drilled for the exit of moisture is inserted into the crucible and the crucible is calcined under working current to sinter it. Sometimes crucibles up to 300 kg in capacity are moulded, dried and calcined separately, outside of the furnace. An acid lining of crucibles can endure an average of 100-150 heats and more. The life of the basic lining is shorter (10-100 heats). The optimum composition of the mixtures used for crucibles has not yet been found and continues to be the subject of research.

The crucibles are destroyed by alternate expansion and compression of the grains of the principal substance in the mixture and the formation of deep spalling as can be seen from the diagram of expansion of acid material (Fig. 144). An outer layer of unsintered compound about 0.5-0.7 of the thickness of the crucible wall should be left to provide for the expansion of the walls. The induction coil is coated with a layer of insulating material.

For acid lining use is made of ground quartzite, to which Belgian foundrymen add 3-4 per cent of clay and about 1 per cent of burnt lime. Then, 2 per cent of soluble glass and 1 per cent of water are added to the dry mixture. Crucibles 100 kg in capacity can endure about 300 heats and 150-kg crucibles about 100 heats.

Burnt magnesite is used as the basic lining. The Czechoslovak engineers specify the chemical composition of the mixture as follows: 2-4 per cent SiO_2; 3.5-5.5 per cent Fe_2O_3; 1.5-2 per cent Al_2O_3; 1.5-8.0 per cent CaO; 80-90 per cent MgO. Use can be made of magnesite brickbat cleaned thoroughly from slag and iron droplets. The latter are removed by means of magnetic separators. The brickbat is ground by millstones and the dust is sifted away. Grains 2-6 mm in size are selected for the mixture.

Soviet plants use chrome-magnesite brickbat. The granular composition of the ground mixture is as follows: grains of size 2-4 mm—15 per cent, 1-2 mm—35 per cent and less than 1 mm—50 per cent. Two per cent of ground fluor spar and 2 per cent of water are added to bind the compound. The compound is mixed and placed into an iron box. The box is covered with a damp cloth and mixture is left there for 48 hours.
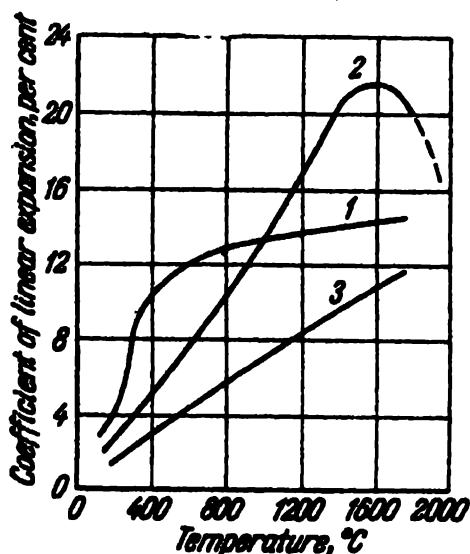
Fig. 144. Change in the coefficient of linear expansion of the materials intended for lining the crucibles of induction furnaces as a function of temperature:

1—acid; 2—basic; 3—ideal lining

Before ramming in, the turns of the induction coil are cleaned, blown over and insulated with a layer 5-6 mm thick composed of four parts of silicon flour and one part of plaster of Paris. The moulding is done with rammers using a sweep template. The coat is dried either by gas burners or electric resistance furnaces. The ready crucible is dried and sintered.

The furnaces where the crucible lining is basic are used to melt alloys very high in manganese, nickel, cobalt, titanium and aluminium including corrosion-proof and heat-resistant steels. In acid crucibles the quality of these steels is impaired.

The melting of steel in coreless furnaces of foundries which make castings from high-alloy grades of stainless, corrosion-proof and heat-resistant steels takes very little time. Taps follow each

other at shorter intervals than from other furnaces, except for the converters. This is an extremely convenient feature for conveyor-type foundries and for automatic moulding and pouring production lines.

The tapping of steel from induction furnaces can be accelerated by combining them with other steel making units as, for example, with an acid converter and even with an open-hearth furnace, which will increase the general technical and economic indices.

A plant in Kladno (Czechoslovakia) producing quality steel has an induction furnace in the open-hearth shop which travelled on guide rails over the pit for moulding boxes. Liquid steel is supplied from two neighbouring open-hearth furnaces into the induction furnace in which the preheated ferroalloys are rapidly melted. High-speed steel is tapped from a 300-kg furnace 30-40 minutes after the liquid steel has been poured in. The furnace operated on the power frequency current—50 cps.

Despite their simple electrical equipment, the furnaces of such low frequency (small storage batteries of the capacitors) are used rather seldom because the charge placed in the crucible should consist of whole and large metal pieces, for example, pieces of axles placed vertically.

In essence, the melting of steel in induction furnaces differs but little from the process observed in electric arc furnaces if we disregard the rate of the process. According to A. Samarin, these reactions proceed at a rate several times faster and require a more accurate selection of the charge since the express-analyses lag behind in time and foundrymen have no time left to correct the errors at the end of the heat before tapping. For this reason, every effort should be made to increase the accuracy of the preliminary analyses of the charge materials and their weighing.

The steel boil in the crucible should be avoided as much as possible when phosphorus need not be removed. No thick layer of slag should be formed, except when absolutely necessary, because it dissolves the lining of the crucibles and puts them out of commission prematurely. Additions to the crucible should be preheated by the heat from the gas burners or special resistance furnaces and the charge should be carefully settled down in the crucible during melting and formation of the bridges.

These methods of deoxidising the metal in steelmaking units and its consequences and the subsequent secondary oxidation of steel give rise to oxide and sulphide inclusions which contaminate the metal and deteriorate its microphysical homogeneity and ductility. The residual amounts of gases in the steel (especially of hydrogen and nitrogen) likewise disturb the microstructure of the cast metal.

The purity of the alloys free from gaseous and nonmetallic inclusions of endogenic type, especially in castings made from high-strength steel, is improved today by melting and pouring the metal in vacuum. In industrial coreless furnaces the vacuum may reach 0.1 mm Hg.

45. COMBINED PROCESSES

Each melting method in a corresponding steelmaking unit has its own merits and demerits. The proper selection of the melting process for the given type of foundry production should be preceded by a thoughtful preliminary technical and economic analysis of the local conditions of production depending, firstly, on the local sources of raw materials, fuel and electric power resources of the given economic zone and, secondly, on the aims of foundry production, quality and quantity of castings, their purpose, by-piece weight, size, number of types, the degree of mechanisation of production and the type of the alloy to be used. In other words, in selecting steelmaking units every effort should be made to achieve the aim set with the minimum input of labour and materials.

The actual economy in obtaining liquid steel should be based on production intensity since this essentially reduces the non-recoverable wastes of heat proportional to the fourth power of the difference between the temperatures of the metal and the environment, and proportional to the melting time. The limits of the highest temperature of molten steel are determined by the production assignment and are constant magnitudes. The duration of heat depends on the selected production process; therefore, in selecting furnaces preference should be given to such furnaces which will operate intensively and produce the metal of the proper composition and temperature within the shortest time possible.

Chemical transformations take a more favourable course in converters and open-hearth furnaces from an economic point of view, while the heat processes are more effective in cupolas and in arc or induction furnaces.

Viewed generally, the melting process in a foundry consists of fusing hard charge materials, preliminary chemical treatment of the liquid alloy based on the oxidation of the carbon, phosphorus and sulphur, and bringing the liquid metal to the specified chemical composition, the process which includes reduction, removal of gases and nonmetallic inclusions and, finally, raising the metal to the required temperature.

During the last decades the world foundry practice has developed various combinations of the work of several furnaces.

Such combined processes have earned the name of duplex and triplex processes when the metal is being melted by two or three different furnaces. In a number of cases these progressive methods proved their indisputable worth.

The side-blown and Thomas processes are the oldest duplex processes which make use of a cupola and a converter or a blast furnace and a converter. From the metallurgical standpoint, the cupola in this process is a thermal unit (heating and melting a hard charge) and the converter is a chemical unit in which the liquid metal is subjected to conversion. An essential shortcoming of this process, as used formerly, is a considerable loss of iron and contamination of steel with gases and nonmetallic inclusions.

To modify radically the duplex process—cupola+converter—the recent practice is to enrich the blast with oxygen or conduct the entire melting process on oxygen alone to obtain steel without nitrogen. Steel can be also treated in a converter body with a basic lining with the aid of a vacuum unit. Before tapping the steel, the tuyere and the mouth are sealed hermetically. A vacuum pump is connected to the shut-off melting chamber and the gases are sucked off; the oxides of silicon, manganese, iron and even of aluminium are reduced. The refined steel becomes by no means inferior to the steel obtained both in an open-hearth and in an electric furnace. The new process can be represented as follows: melting in a cupola with a basic lining with hot blast, blowing pig iron with oxygen in the basic converter until all carbon is burned and feeding subsequently liquid pig iron at the end of the heat to deoxidise with carbon and to carburise the steel, remove the slag and treat the metal in vacuum before tapping it from the converter.

High-quality low-alloy structural steel may also be tapped by this method.

This process can be recommended for the areas where the cost of electric power is rather high.

In the areas with oil and gas deposits some of the coke in the cupola should be replaced by black oil or highly calorific gas.

English foundrymen tried a new duplex process: cupola+ +open-hearth furnace. Almost all or all hard metallic part of the charge is melted in a hot-blast cupola with a dolomite lining and the liquid metal is transferred to an open-hearth furnace. In the bath of the open-hearth furnace the admixtures are oxidised, part of the phosphorus and sulphur are transferred to slag and the melt is refined. The output of the open-hearth process is increased almost twofold since the work of heat is transferred to the cupola, a more economical unit, when the hard charge has melted down. The shortcoming of this process is the need for extra coke. This process has not found application in the Soviet Union. The increased output of the open-hearth furnace can be

attributed to the fact that almost half of the time is expended in it to melt down the hard charge. With this duplex process the unproductive consumption of fuel and time is obviated.

In the early 20th century Kranz' triplex process — cupola + + converter + electric arc furnace — came into use in the U.S.A. The refined metal is supplied from a basic electric furnace for pouring (about 50 per cent of the charge weight). Then the low-carbon steel from the converter and part of the liquid pig iron from the cupola are poured into the furnace to carburise the metal. The mixed metal is held in the arc furnace, refined and delivered for pouring, etc. The shortcoming of this process is that the steel-makers are occupied at the three units which operate simultaneously. This process is not used in the Soviet foundries.

The economically justifiable duplex process is an arc or cruci-ble furnace + a furnace for heating and sometimes for melting ferroalloys. The tapping rate is increased, especially when high-alloy steels, such as Г13Л, are produced.

In conclusion we shall point out certain basic requirements which should be complied with in selecting a steel smelting method. The priority is given to the main aims of foundry produc-tion—the quantity and quality of castings.

In the choice of a steel foundry process the main prerequisite is the serial aspect of production and its synchronism. For exam-ple, conveyor-type foundries must ensure uninterrupted supply of ready moulds and continuous pouring. This condition can be met by such furnaces from which ready heats are tapped at small intervals. A pouring conveyor can move continuously if the heat cycle does not exceed the duration of pouring of one heat, not more than 30-40 minutes, approximately. This requirement is met by a series of several arc or induction furnaces which pro-duce steel of one grade. A continuous pouring of steel of two grades requires obviously two conveyors and two times more furnaces or else the work on one conveyor should be rearranged and the groups of moulds filled now with one and now with the other grade. The steel may be obtained in a cupola and a converter while the final melting should be performed in induction bale-out pot furnaces.

In small-lot production other factors grow in importance: such as by-piece weight of the castings, wall thickness, number of steel grades, etc. The metal for thin-walled castings should be poured at a higher temperature of the metal. In this case, prefer-ence should be given to a converter which can ensure maximum temperature without the hazard of igniting the roof of the melting chamber. When steel scrap is scarce the metal should be prefer-ably melted in a converter if the high melting loss of iron amount-ing to 12-15 per cent and to 8-12 per cent with an oxygen blast may be neglected. The second place belongs to the induction

furnaces. Induction furnaces with a charge capacity of 18 tons have been designed in Sweden. The third place is occupied by electric arc furnaces and the fourth by open-hearth furnaces.

Engineering plants ordinarily have enough home steel scrap more or less free from sulphur and phosphorus. This explains a widespread use of acid electric arc furnaces in steel foundries. The furnaces operate on the principle of nonoxidising remelting of the scrap. If there are several acid arc furnaces in the shop provision should also be made for furnaces with a basic lining for producing chromium-containing steel, as well as for a periodic refining of the home scrap which gets gradually enriched with residual gases and, especially, with nitrogen and hydrogen. The hydrogen can be got rid of mainly during the period of vigorous boiling of the bath in the oxidation period of melting.

When heavy steel castings are produced in the areas where the cost of electric power is too high, good economic effect can be obtained from the use of open-hearth furnaces which have become classical smelting units at heavy engineering plants.

Open-hearth furnaces at engineering plants operate only on hard charge. The chips obtained in large amounts should be made into briquettes and remelted at the plant. If the pig iron is scarce and has to be hauled from afar it will be good practice to install cupolas near the open-hearth furnaces and use the duplex process—cupola + open-hearth furnace—in order to raise carbon content in the cupola to the required level before the molten metal is delivered to the open-hearth furnace.

In areas with cheap sources of electric power it is irrational to erect open-hearth furnaces even at heavy engineering plants. It is more profitable to install several large electric arc furnaces with the charge capacity of 180 tons and more.

When a broad variety of alloy steel grades is available electric arc furnaces in steel foundries have time and again proved much superior to open-hearth furnaces.

If there is an abundance of cheap oxygen, the oxygen blast in arc furnaces with a basic lining can raise the efficiency of electric melting still further as compared with the open-hearth process.

46. VACUUM TREATMENT OF STEEL

There is not the least doubt as to the role the gases and nonmetallic inclusions play in decreasing the macro- and microphysical homogeneity of steel and, hence, in reducing its ductility and impact strength, in deteriorating the heat resistance and the magnetic properties of alloys. The treatment of liquid steel in

vacuum ranks among the latest achievements in the theory of deoxidising the steel and cleaning it from harmful admixtures.

The vacuum makes it possible to organise effectively the production of many metals and alloys free from such admixtures as carbon, oxygen, nitrogen and hydrogen. It is only with the aid of vacuum that the advantages of carbon and hydrogen as deoxidising agents over other elements can be realised.

The interaction of carbon with oxygen dissolved in a liquid metal causes the formation of gaseous reaction products—carbon monoxide and carbon dioxide. In this case deoxidation is not accompanied by the formation of oxides similar to the oxides of silica or manganese which do not dissolve in metal. But, they may be trapped in metal as dispersed droplets or hard particles since the rate of their removal is too low. Consequently, in deoxidation with carbon in a vacuum we can obtain a metal with very little dissolved oxygen and with very few oxides, the products of the reactions of deoxidation with silicon, manganese, aluminium, etc. When the gaseous carbon monoxide is removed from liquid metal the dissolved hydrogen and nitrogen increase in amount. In vacuum furnaces the carbon monoxide is being continuously removed which increases its deoxidising ability.

The equilibrium constant of the reaction of deoxidation of the carbon dissolved in liquid iron can be determined from the equation

$$K' = \frac{[\% \, C] \, [\% \, O]}{P_{CO}} \, . \tag{209}$$

When the pressure of the carbon monoxide is 1 atm at 1600°C, the value of K' changes from 0.0020 to 0.0025. When the pressure of the carbon monoxide decreases the value of the constant K' remains the same if the product [%C] [%O] is correspondingly reduced. Fig. 145 shows the equilibrium concentrations of oxygen and carbon dissolved in liquid iron at 1600°C as a function of the pressure of carbon monoxide. In other words, the diagram shows the deoxidising ability of carbon at the carbon monoxide pressure equal to 1.0 and 0.1 atm; 10 and 1 tores (1 tore=1 mm Hg; 1 m Hg=1,000 tores).

For the sake of comparison we give deoxidising abilities of manganese, silicon and aluminium at the same temperature. By decreasing the pressure the carbon can be made a much stronger deoxidiser than silicon and aluminium. The advantage of deoxidising with carbon lies in the fact that the product of the reaction is the volatile compound—carbon monoxide and not silica, alumina and other oxides which are very difficult to remove from the deoxidised metal to improve its impact strength and the physical homogeneity of the alloy.

Ten years ago, vacuum induction furnaces were a rarety at iron and steel plants and the output of vacuum-treated steel was a bare tens of kilograms. Today, vacuum metallurgy is making gigantic strides and the yield of such steel runs into thousands of tons.

The most essential advantage of vacuum treatment is the decrease in the carbon content and also in obtaining low-carbon
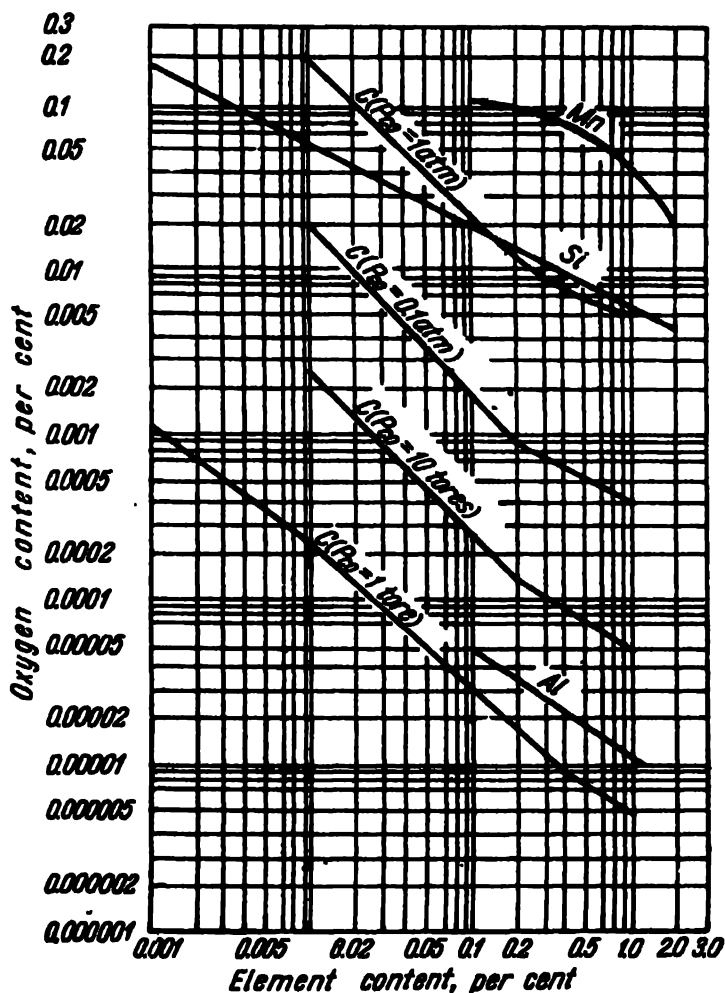
Fig. 145. Change in the equilibrium concentrations of oxygen and carbon dissolved in liquid iron and the content of the deoxidising elements depending on the partial pressure of carbon monoxide (Samarin)

grades of steel, for example, transformer or heat-resistant chromium steel. It is extremely difficult to remove the last tenth and hundredth fractions of per cent of carbon in final melting of mild steel. The treatment in vacuum makes is possible to obtain steel with 0.03 and even 0.01 per cent C, which is extremely vital in producing shaped castings.

The characteristic change in the composition of the transformer steel after vacuum treatment in an induction furnace is. represented in Table 57.

Table 57

**Chemical Analysis of Transformer Steel After Vacuum
Treatment in an Induction Furnace, per cent**

(Samarin)

Furnace	Heat No.	C	Si	Mn	P	S
Ordinary . . .	1	0.040	3.50	0.15	0.005	0.017
	2	0.040	3.53	0.13	0.013	0.027
	3	0.030	3.24	0.15	0.012	0.018
	4	0.030	2.75	0.10	0.007	0.013
	5	0.025	2.63	0.10	0.008	0.018
Vacuum . . .	6	0.015	3.12	0.13	0.023	0.036
	7	0.020	2.92	0.15	0.035	0.036
	8	0.020	3.35	0.14	0.026	0.037
	9	0.020	3.53	0.15	0.023	0.033

Table 58

**Content of Oxygen, Hydrogen, Nitrogen and Nonmetallic
Inclusions in Steel, per cent**

(Samarin)

Furnace	Heat No.	O	H	N	Nonmetallic Inclusions
Ordinary . .	1	0.0192	0.0002	0.0056	0.039
	2	0.0177	0.0003	0.0030	0.034
	3	0.0195	0.0004	0.0044	0.050
	4	0.0145	0.0003	0.0048	—
	5	0.0182	0.0002	0.0028	0.038
Vacuum . .	6	0.0019	0.00005	0.0028	0.0042
	7	0.0026	0.00005	0.0042	0.0070
	8	0.0022	0.00005	0.0037	0.0050
	9	0.0025	0.00005	0.0028	0.0030

After vacuum treatment in a furnace the carbon content decreases but the amount of phosphorus and sulphur remains high since no slag is present during the vacuum treatment. The watt losses decrease after the vacuum treatment by 12-25 per cent due, apparently, to a smaller content of gases and nonmetallic inclusions (Table 58).

During the vacuum treatment the amount of nitrogen remains practically the same, but there is a sharp decrease in the content of hydrogen and oxygen and nonmetallic inclusions.

It is extremely important to reduce the carbon content in alloys with chromium. This can be easily seen from the equilibrium constant of the reaction

$$CrO_{sol} + [C] = [Cr] + CO_2; \tag{210}$$

$$K = \frac{P_{CO}\,[\%\ Cr]}{[\%\ C]}; \tag{211}$$

$$\lg K = -\frac{15,200}{T} + 9.46. \tag{212}$$

The calculation data are indicated in the diagram in Fig. 146. The Lectromelt Co. receives its ferrochromium in vacuum furnaces by decarburising high-carbon ferrochromium by means of
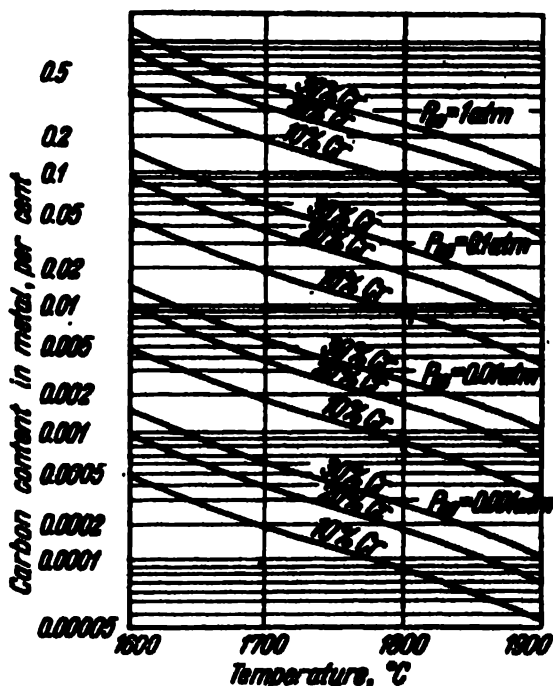
Fig. 146. Calculated dependence of the equilibrium contents of carbon in chromium steel on the temperature and the partial pressure of the carbon monoxide (Samarin)

oxidising the carbon with silica oxygen. For this purpose, briquettes made from a powder of high-carbon ferrochromium and quartzite are held in a vacuum furnace below the melting point. The ferrochromium obtained is very low in carbon; part of the silica is converted into silicon. Besides, the briquettes reveal the remains of silica.

Low-carbon ferrochromium can be used to obtain stainless and heat-resistant steels with a carbon content not over 0.03 per cent and with a standard amount of silicon.

Today, this method is also used to obtain other low-carbon ferroalloys, ferromanganese, for example.

The vacuum treatment opens up new possibilities for improving the production of steel castings. B. Ladyzhensky suggested an interesting method for using a vacuum unit together with a converter. Pig iron or steel scrap is remelted in a cupola with

a basic lining where sulphur changes to slag. The pig iron in the converter is blown with oxygen until the required carbon content is obtained. Then, the oxygen tuyere is cut off and the slag is removed. The mouth is shut off with a lock to which a vacuum reservoir is connected. The steel is deoxidised by the remains of carbon and is cleaned from gases and nonmetallic inclusions. Such units have not found so far practical application but they are extremely promising, especially for obtaining critical castings made from carbon and alloy steel. Alloying can be also conducted in the ladle without preheating the ferroalloys.

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